NOVEMBER, 1958

JOURNAL OF THE IRON AND STEEL INSTITUTE

4 GROSVENOR GARDENS · LONDON SWI

Established 1869 · Incorporated by Royal Charter 1899

VOL. 190 PART 3

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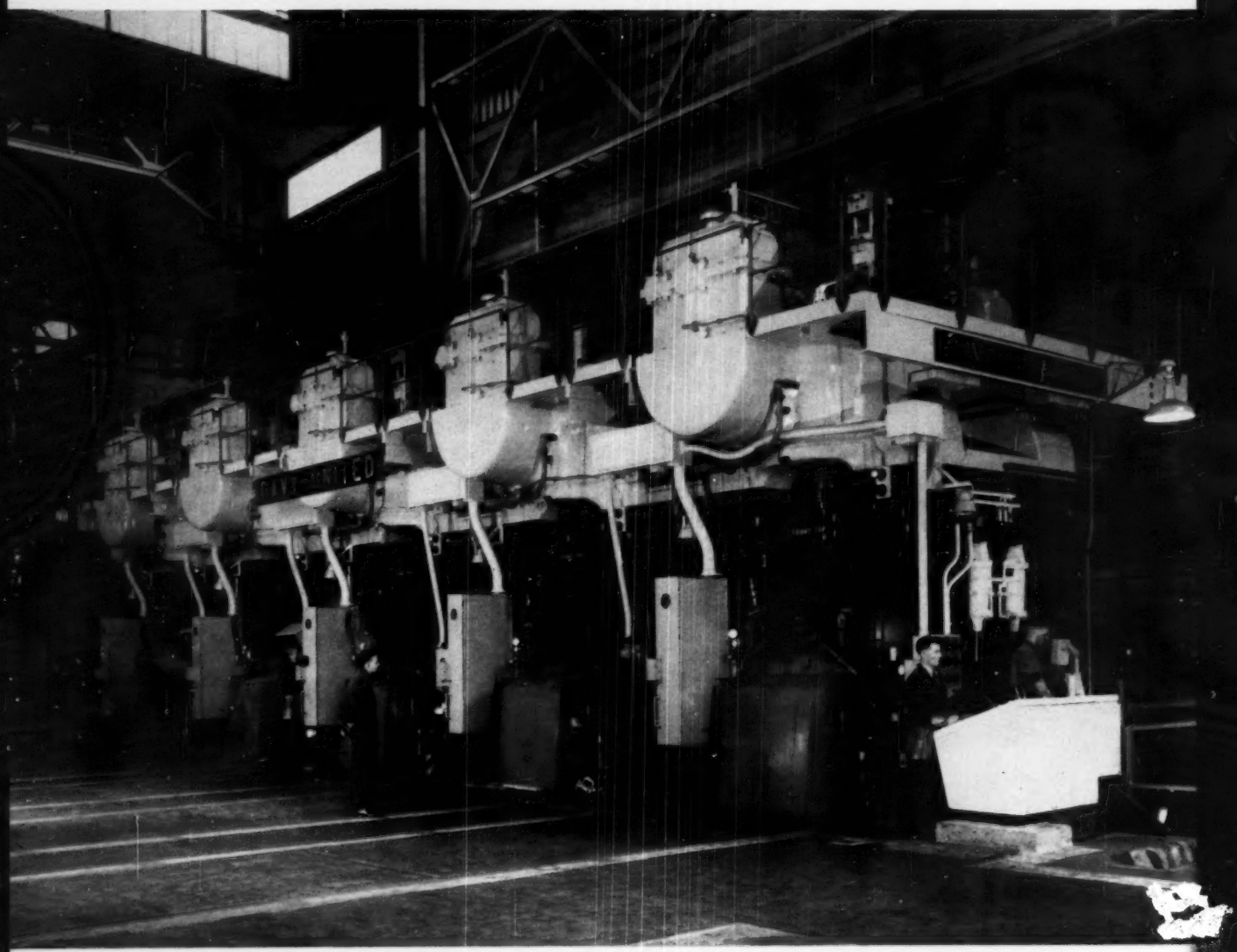
more strip for **TINPLATE**



The 21″ and 53″ × 42″ 5-Stand Tandem Cold Strip Mill at Velindre, rolling steel strip for tinplate at 5,000 feet per minute.

"The Velindre Tinplate Mill of the Steel Company of Wales is the most modern in the world and has come into operation more quickly and smoothly than was to have been anticipated."

The Chairman, The Metal Box Company Limited, in Statement to Shareholders 21st July, 1958.

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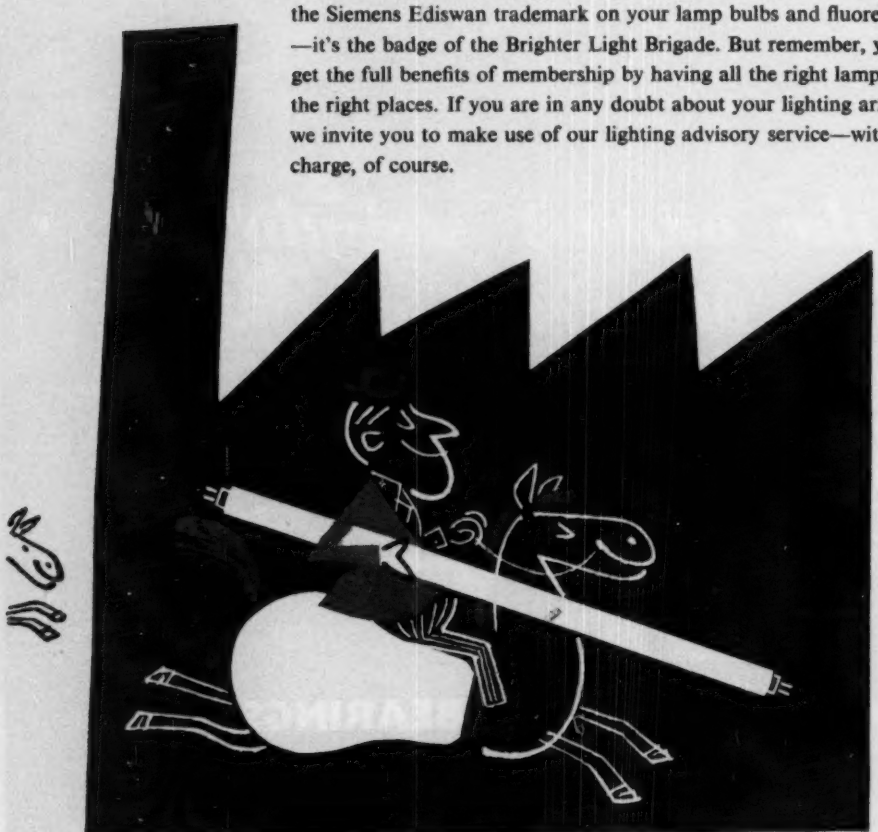
Have you got the right lamps— in the right places?

In the very wide range of Siemens Ediswan lamps there is a 'best' type, size and power of lamp for every place and every purpose. It usually costs no more—and may often cost less—to fit the correct Siemens Ediswan lamp than to put up with a lamp or a fitting that is 'near enough'. Look for the Siemens Ediswan trademark on your lamp bulbs and fluorescent tubes —it's the badge of the Brighter Light Brigade. But remember, you only get the full benefits of membership by having all the right lamps in the right places. If you are in any doubt about your lighting arrangements, we invite you to make use of our lighting advisory service—without charge, of course.



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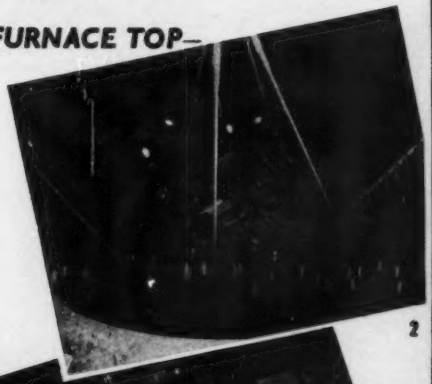

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—TO FURNACE BOTTOM

1 *'Lumpy going'—half-way down hearth.*
2 *Men removing throat armour.*
3 *Cat Traxcavator commencing digging, Cast House Floor Level.*
4 *The two 'cats' meet—final 'break through'.*

1

Scenes of operations at No. 1 Blast Furnace
(photos by permission of John Summers & Sons Ltd., Shotton)

Furnace blown out	23rd April 1958
Demolition commenced	26th April 1958
Demolition completed	13th May 1958

TOTAL TIME wrecking burden, lining, hearth and bear
18 DAYS
(contract carried out in association with Salem Engineering Co. Ltd.)

Your enquiries are invited.

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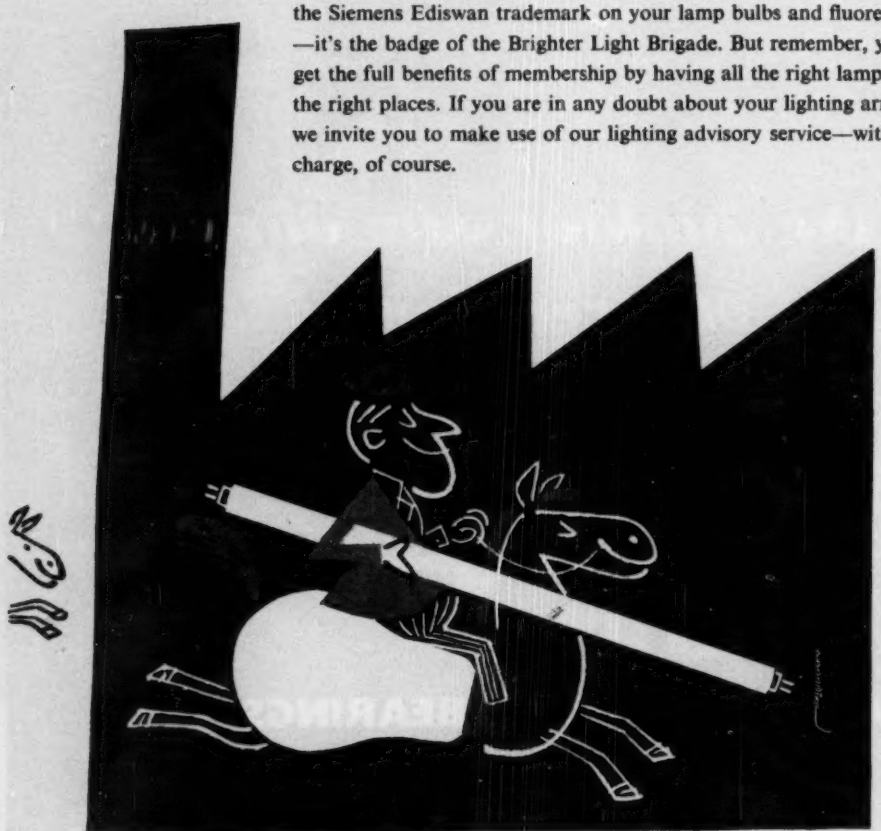

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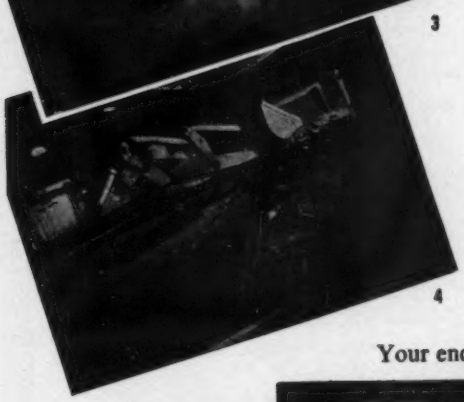

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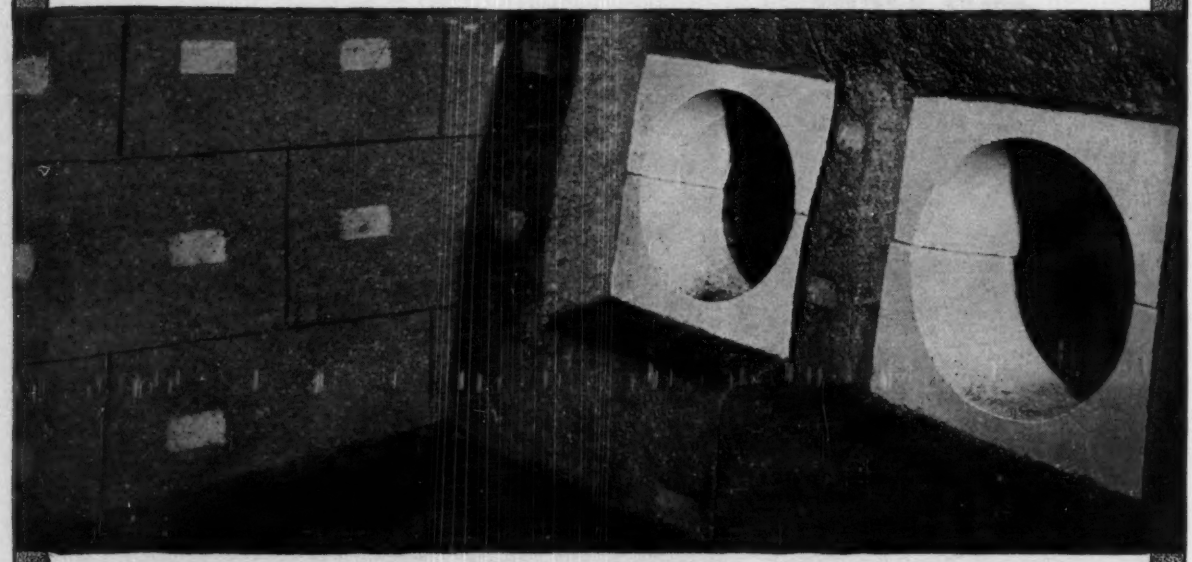

Front and side walls of a Queen Mary boiler in Tri-Mor High Temperature Mouldable, with MR 60 anchors. Burner quarls in Tri-Mor High Temperature Castable.

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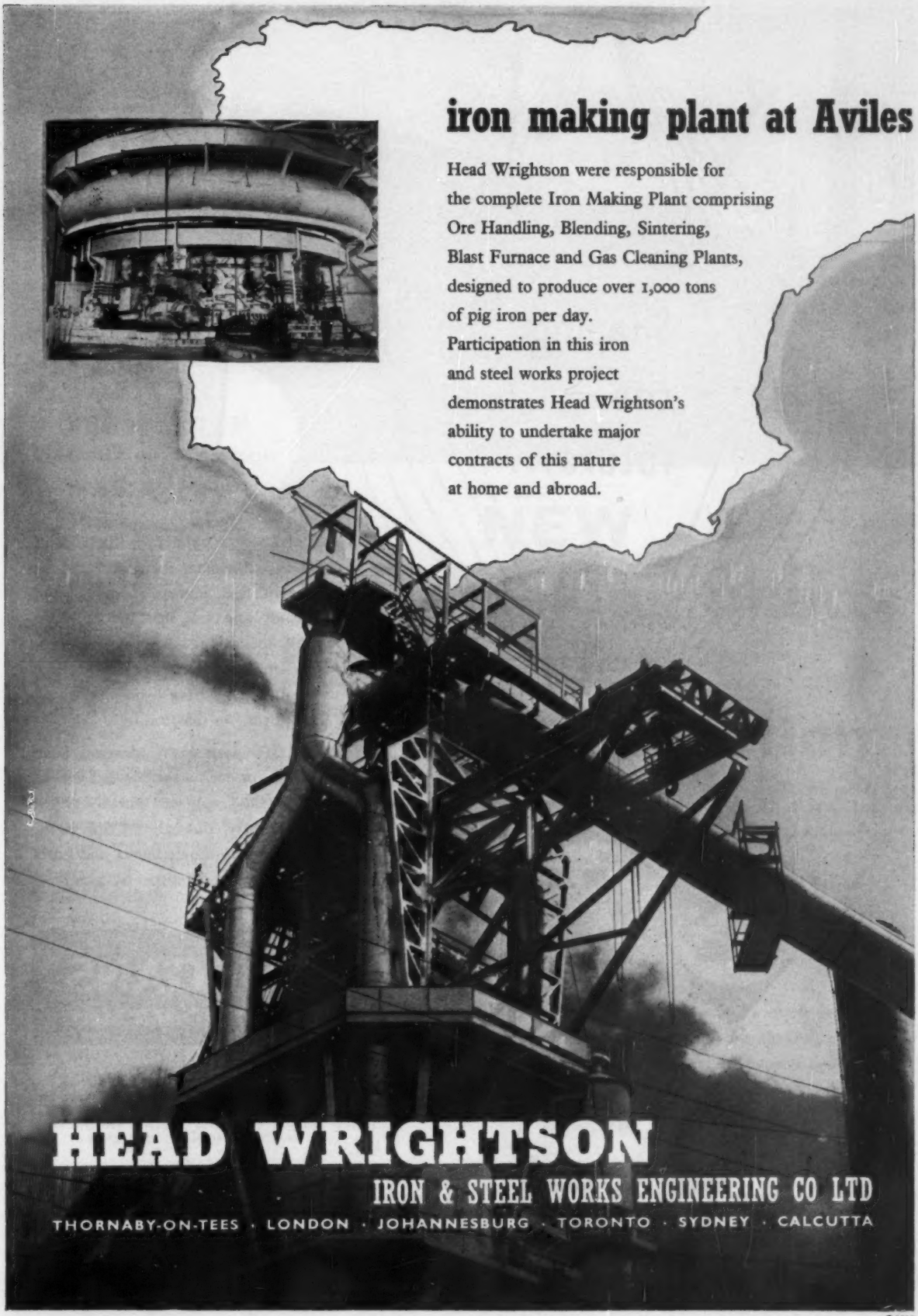
iron making plant at Aviles
Head Wrightson were responsible for the complete Iron Making Plant comprising Ore Handling, Blending, Sintering, Blast Furnace and Gas Cleaning Plants, designed to produce over 1,000 tons of pig iron per day.
Participation in this iron and steel works project demonstrates Head Wrightson's ability to undertake major contracts of this nature at home and abroad.
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IRON & STEEL WORKS ENGINEERING CO LTD
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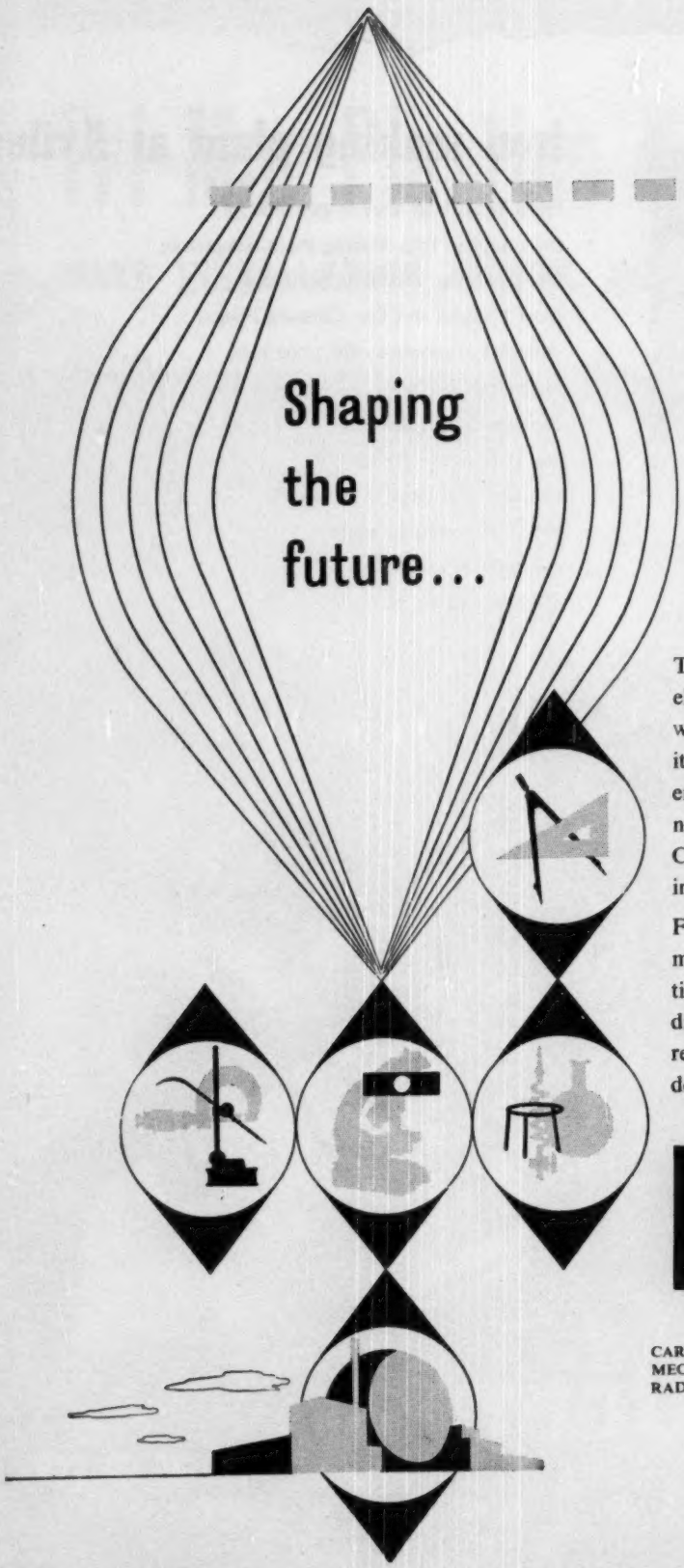

There is already a pointer, in the fields of electronics and nuclear engineering, of the way the industrial future is shaping. And it is already apparent that the chemists, engineers, metallurgists, physicists, technologists and craftsmen of The Morgan Crucible Company have a big and growing part to play in that future.

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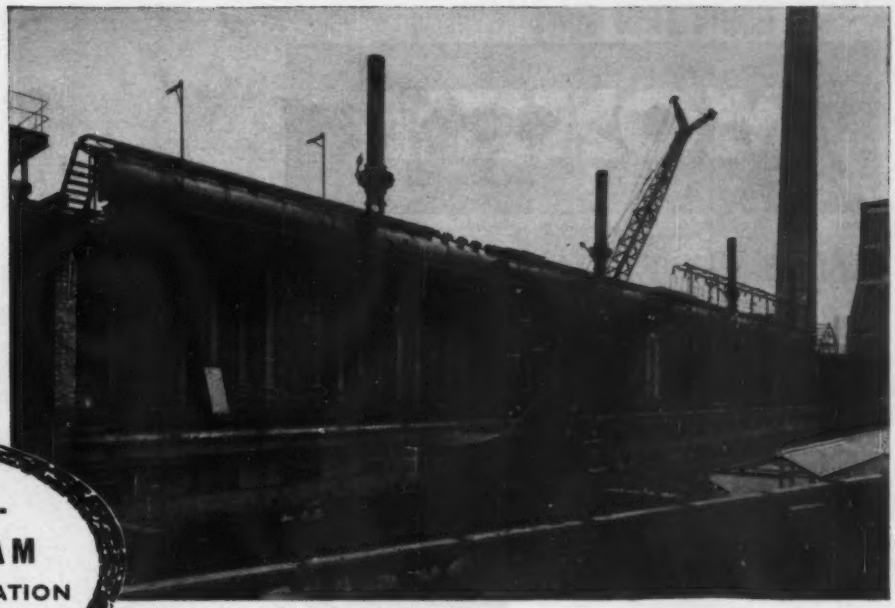

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FIFTH INSTALLATION
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THIRD INSTALLATION
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THIRD INSTALLATION
55 OVENS

NEW coking plants under construction

The batteries of the second installation of W-D Koppers Coke Ovens for the Appleby-Frodingham Steel Company now nearing completion.

The carbonisation capacity of the two new batteries of 33 Ovens is 1330 tons of coal per day. The previous installations of W-D Koppers and Koppers Ovens each have two batteries of 33 Ovens wih similar carbonising capacities.

The latest installation includes a new ovens bunker with additions to the coal and coke handling plants and extensions of the by-product, benzole and boiler plants.

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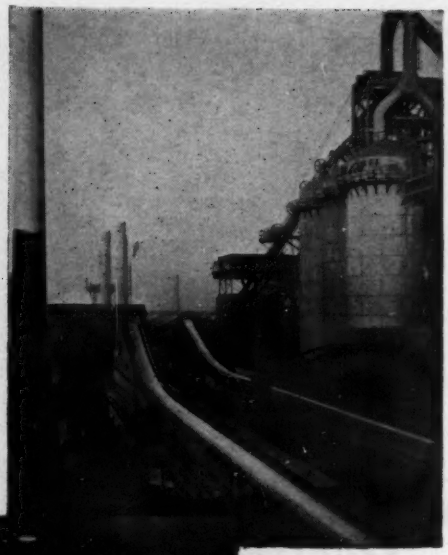

Conveyors Feeding Iron Ore to Blast Furnace Bunkers.

Ore Plants
Fuel Plants
Bulk Handling
Belt Conveyors
Belt Conveyor Idlers
Slat Conveyors
Scraper Conveyors
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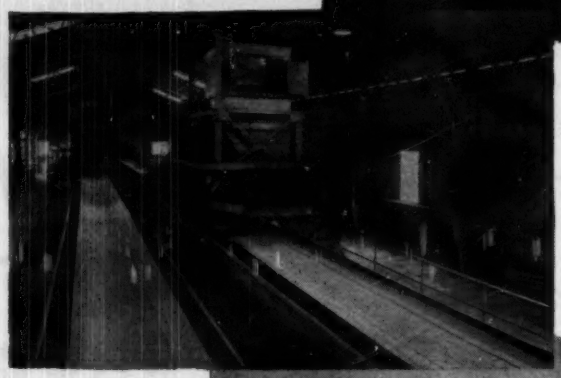
Portable Conveyors
Roller Conveyors
Bucket Elevators
Swing Tray Elevators
Skip Hoists
Vibrating Screens



Trippers feeding Ore to Dockside Bunkers.

Boomstacker stock-piling Iron Ore.

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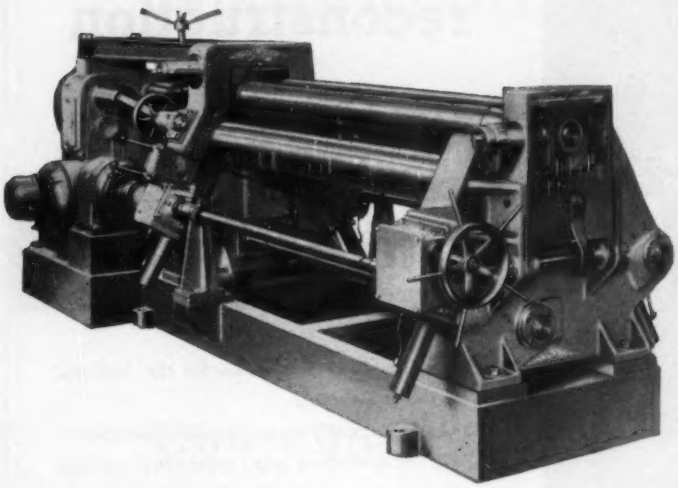

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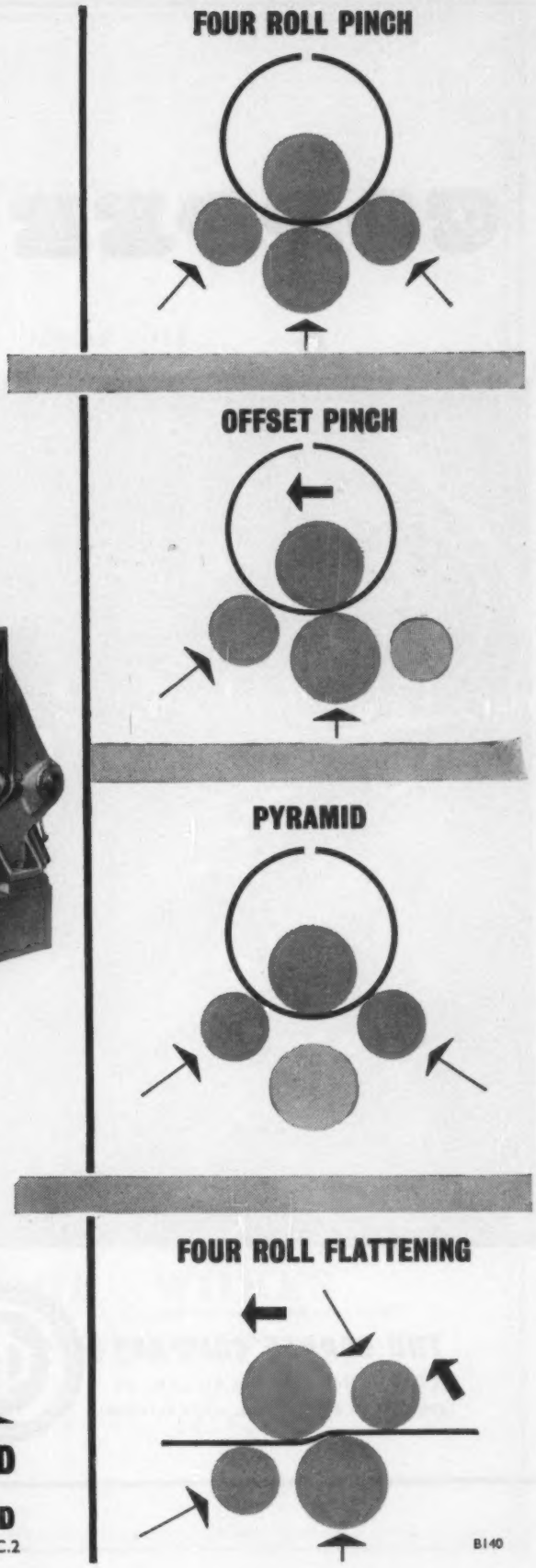

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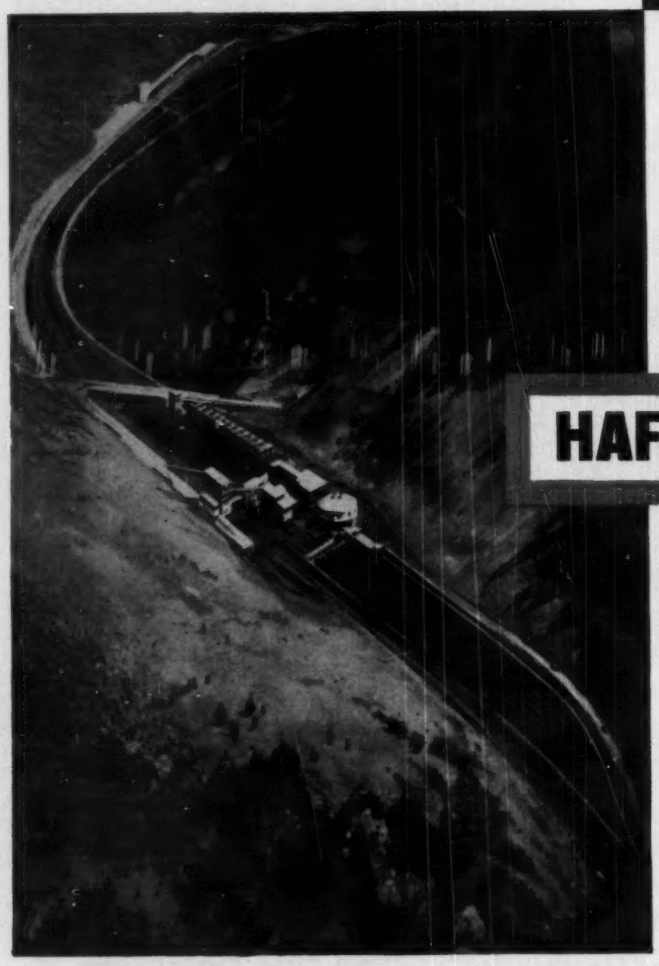

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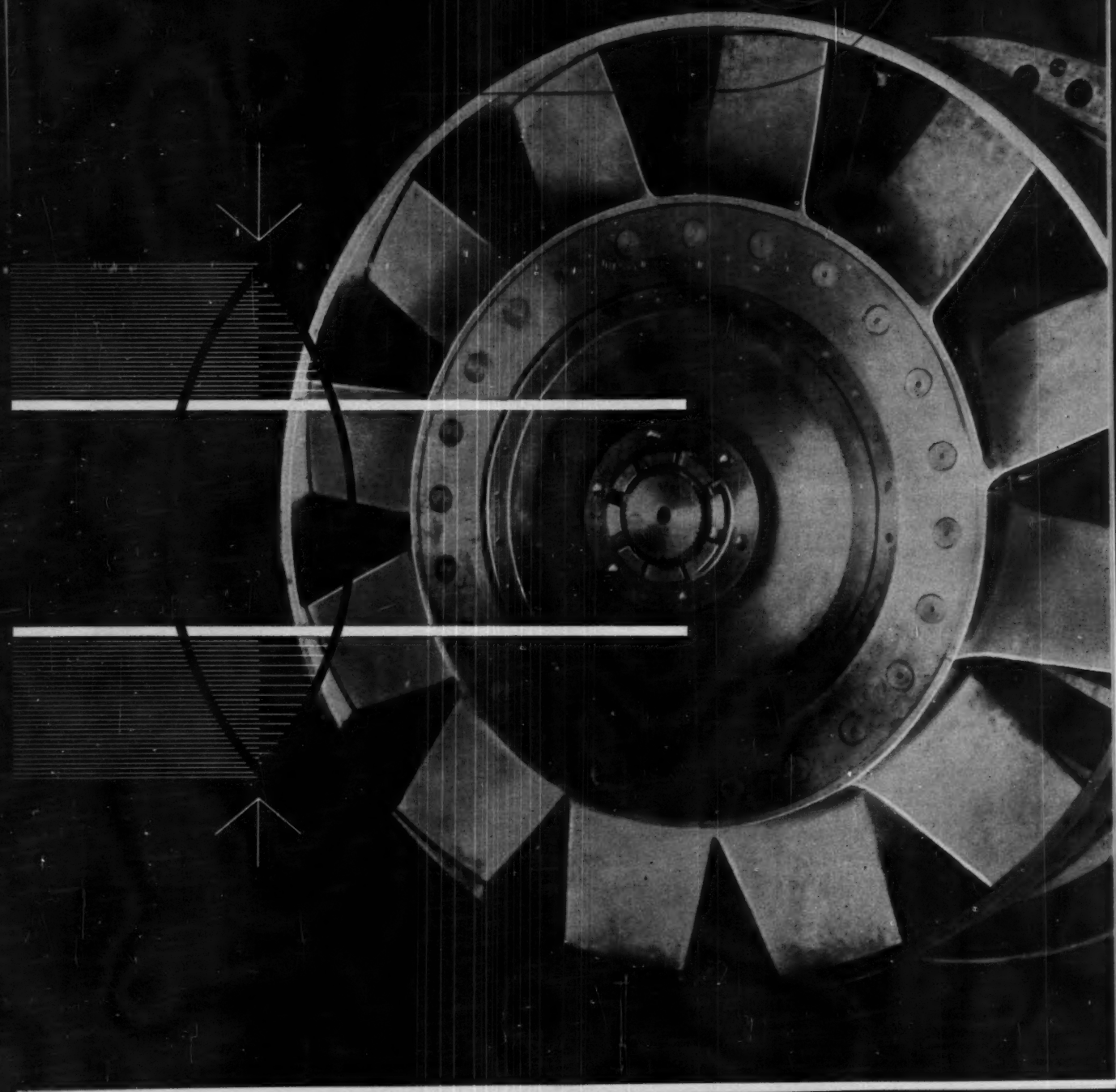
APL blower oil on



sentry-go at Bradwell...

The crucial points of the high pressure CO_2 coolant circuit at Bradwell will be the blower shaft seals. Every minute of every working day Shell APL blower oil will guard these escape points—being continually pumped into the seals and bearings, led away, purified and recirculated.

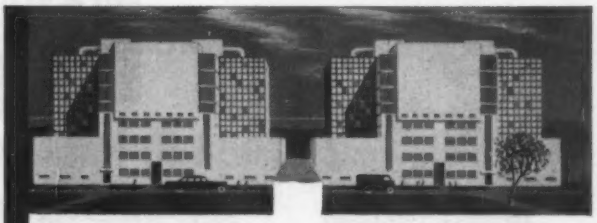
This very remarkable oil presents a two-way seal. Outwards, there is a minimum CO_2 loss as the oil has the lowest possible gas solubility. Inwards, in spite of the heat of the gas, oil vapour contamination is negligible thanks to the extremely low vapour pressure of the oil.

The research that went into APL blower oil is characteristic of the way Shell set about doing things. It was conducted at Shell's Research Centre at Thornton in close collaboration with the U.K. A.E.A. and the blower manufacturers. In the course of much fundamental research, a wide range of oils was subjected to vapour pressure and gas solubility tests in the laboratory. Selected oils from this range were used in the bearings and seals of a blower rig. In 1956, after four years of research, the finished product joined the Shell Atomic Power Lubricants range—marketed under the name of Shell APL 729. This oil has been in use at Calder Hall since the autumn of 1956.

The moral of the story is that Shell research is supremely applicational. The Centre at Thornton is always ready to work with even the most specialised sectors of industry to produce the right oil for the job. If you and your organisation have any major lubrication problem it pays to get in touch with your local distributor of Shell Industrial Lubricants.

The Research Story

Though the sealing action demanded of Shell blower oil is common in other industrial equipment such as hydrogen-cooled alternators, there were quite a few additional problems. It was necessary for the oil to have long life, low vapour pressure, low gas solubility, good thermal stability and high film strength —to be able to withstand high temperatures and to be resistant to all kinds of corrosive influences, including carbonic acid. The crux of the research was to combine all the above requirements into one oil, in order to minimise back diffusion of molecules which would contaminate the reactor.



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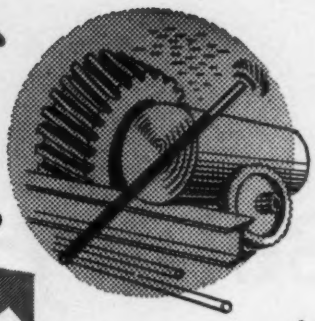
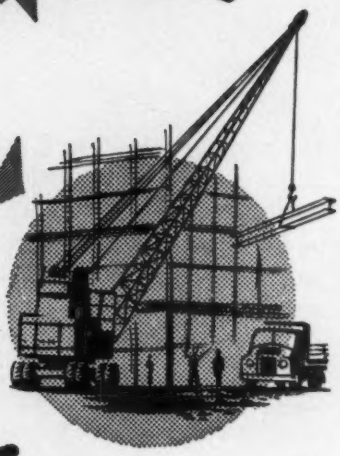


dismantling.....



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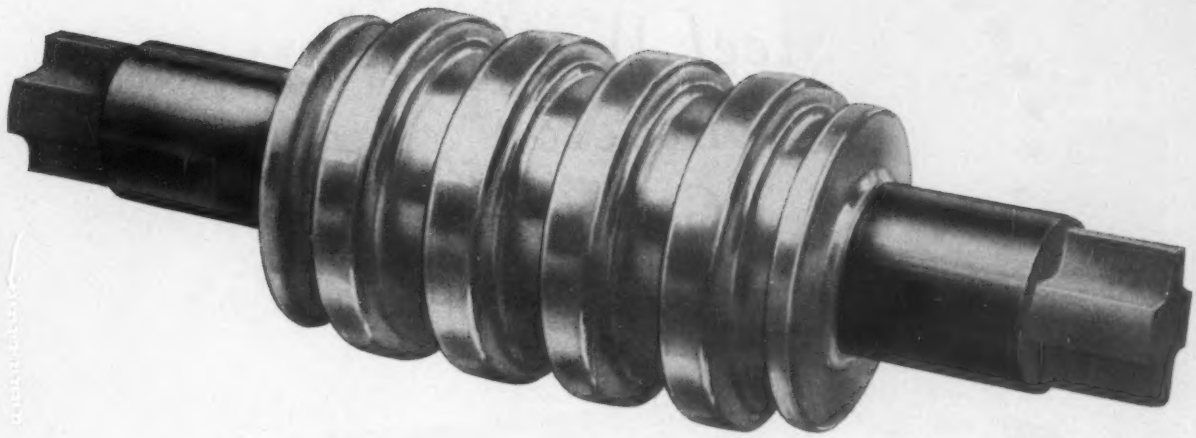

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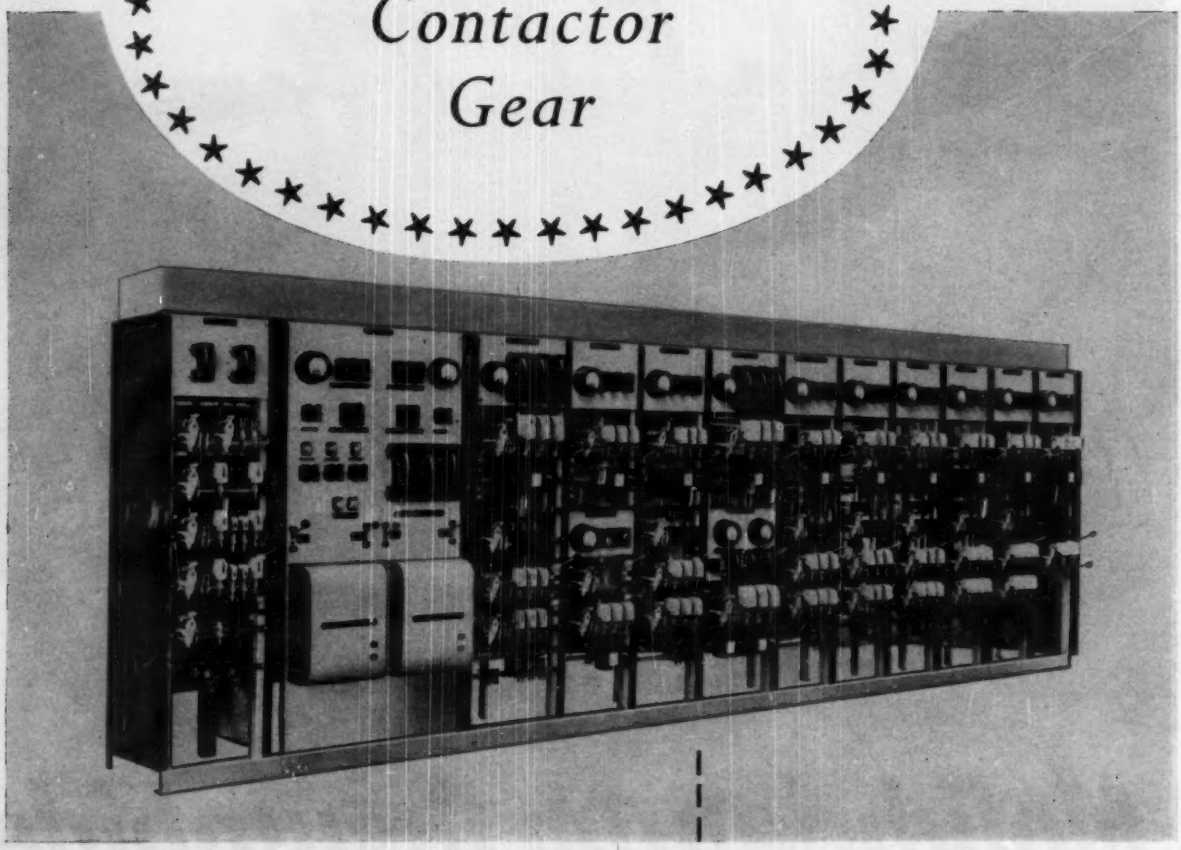

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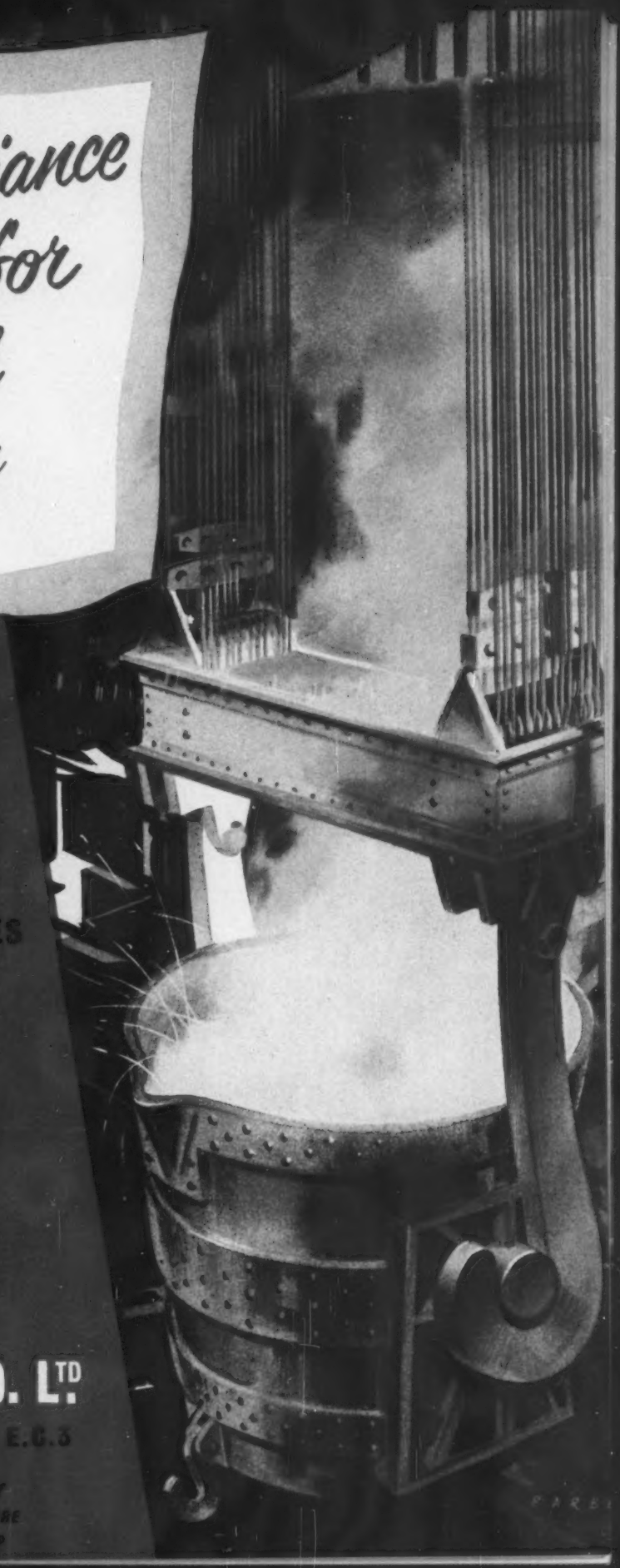

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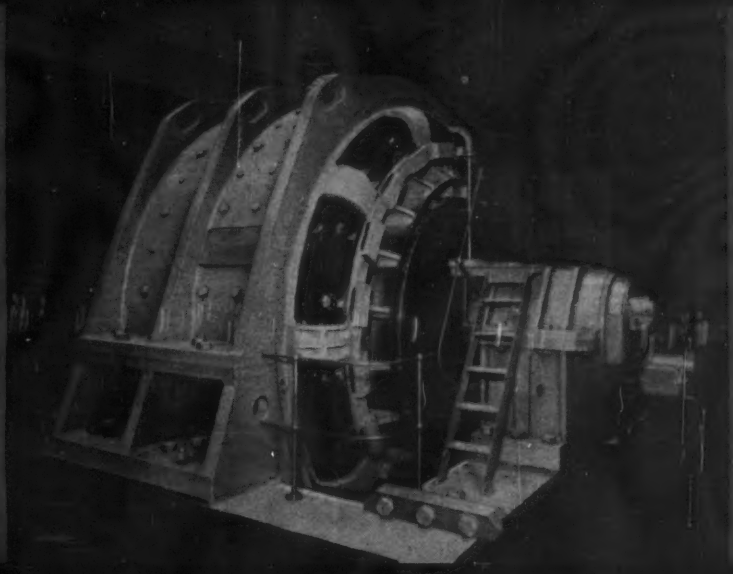
FORWARD

to

REVERSE

in 1½ seconds!

A small lever is moved – and 58 tons of armature reverse. From a base speed of 50 r.p.m. one way to 50 r.p.m. in the opposite direction. In 1½ seconds. This giant G.E.C. 5,000 h.p. motor is driving a cogging mill at the Redbourne Steel Works of Richard Thomas and Baldwins Ltd. It typifies the speed and precision of the electric drive.



Less spectacular than the mammoth motors but equally important to steel production are the many hundred auxiliary drives operating in a modern works where the utmost reliability is essential. Here, again, most of the handling requirements subject the equipment to very severe stresses. G.E.C. motors are as tough as the jobs they're designed to do.

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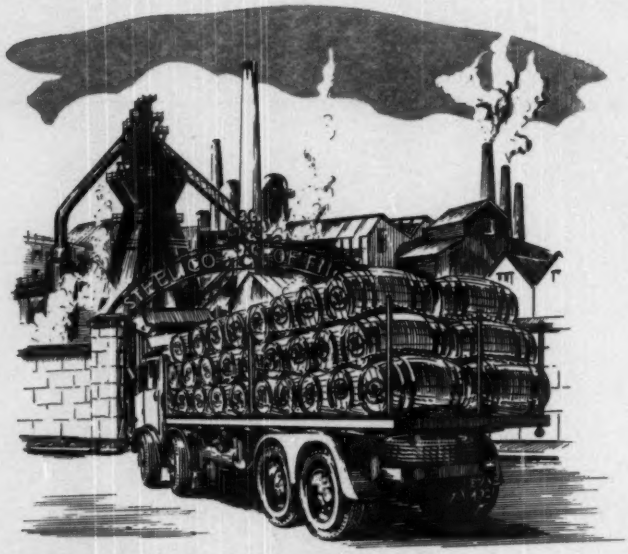

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"St. Lawrence" Ferro Silicon is available for immediate despatch from our many storage depots in the United Kingdom. Whatever your requirements, we can deliver into your works the next day if necessary. This delivery service is backed by our team of experienced field metallurgists who are ready to give technical assistance when required. And bear in mind also that whenever you buy "St. Lawrence" Ferro Silicon you are not only using the finest material available to-day, but also the product of British labour from a British plant. So buy British it's the very best.

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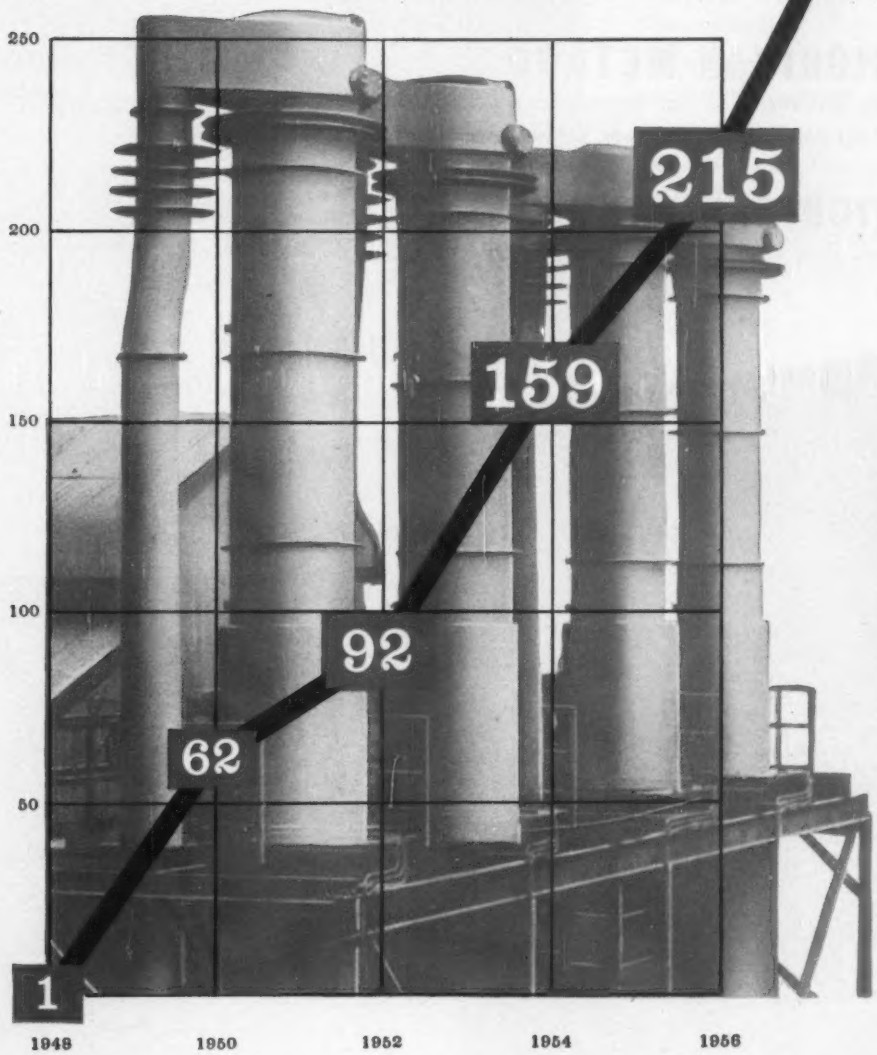

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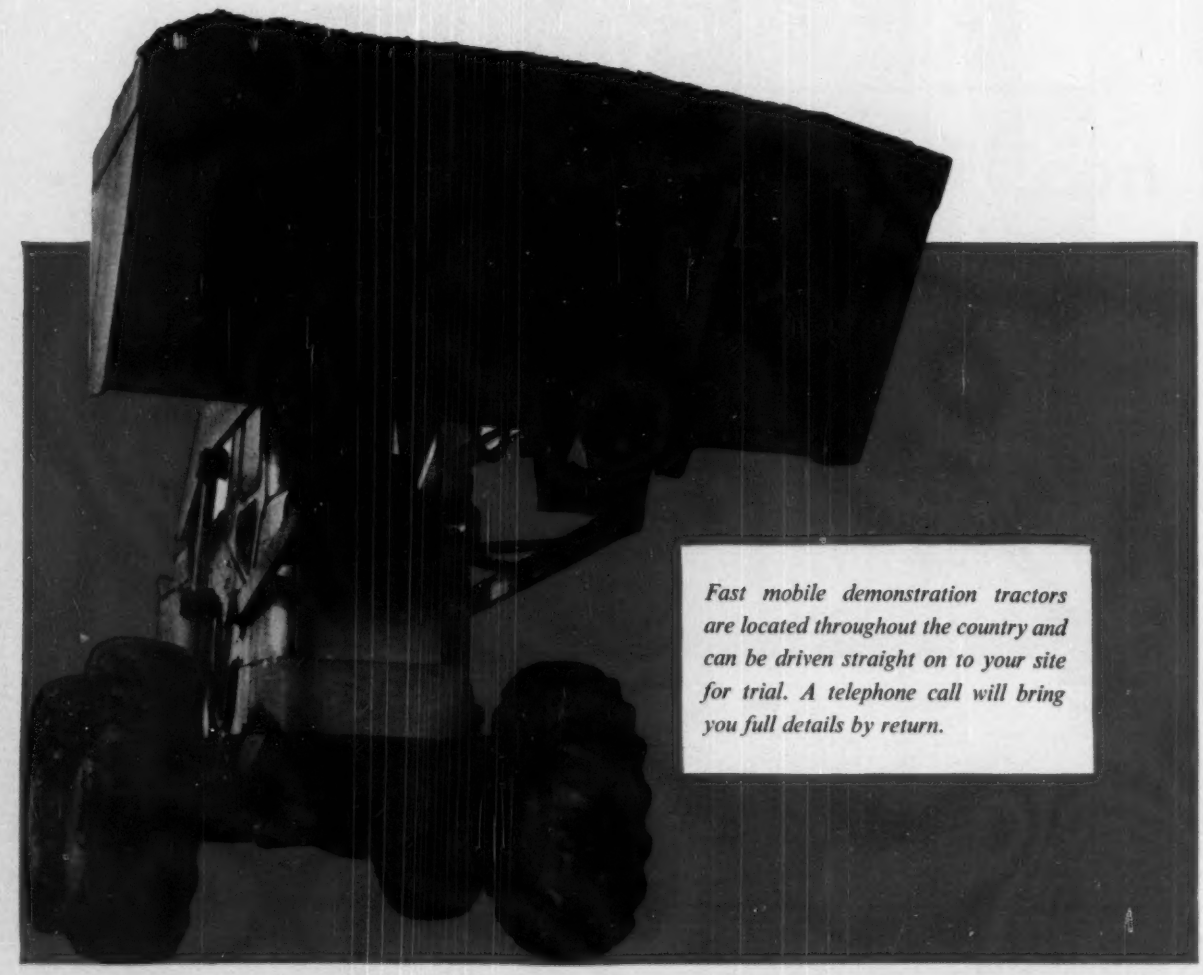

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Electrical Aids in Industry — *Data Sheet* **No.1**

Electro-Heat

The ways in which electricity can be used to advantage in industry are many and varied. Some are well known but others are not known well enough. For this reason the Electrical Development Association has prepared a series of detailed data sheets on various applications which will be printed in this journal from time to time.

This Data Sheet is the first of a number devoted to electro-heat—that is, heat produced by electricity for the processing and treatment of materials. Other uses will be dealt with later on in the series.

There is an unalterable physical law that the efficiency of conversion of *electrical* energy to *heat* energy is 100 per cent. The same cannot be said of the efficiency of combustion of any solid or liquid fuel.

All the applications of electro-heat have these advantages in common :—

1 Electro-heat is clean both in regard to its application and the method of generation.

2 It can easily be controlled more precisely than any other form of heat, manually or automatically.

3 It can be brought to the job instead of having the job brought to it.

4 It permits better use of floor space and the elimination of unnecessary handling.

5 In many of the newer processes it is the only possible form of heat which can be used.

6 It often ensures a higher quality of products with fewer rejects.

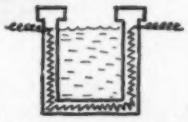
7 It gives the best working conditions.

8 It reduces or eliminates fire and explosion hazards.

Below will be found brief notes on some of the various methods of generating heat by electricity.

Resistance Heating

This is the best known form of electric heating. The elements provide a high resistance to the passage of electricity and thus heat is generated. It can be used in furnaces for melting or heat treatment of any material whether metal or not, or the resistance of the workpiece itself can cause the generation of heat.

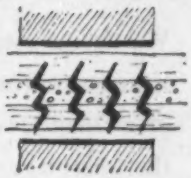


Induction Heating

Eddy currents are induced in the surface of a conducting workpiece, heating it up. The depth to which this heating will penetrate is determined by the time it is given.

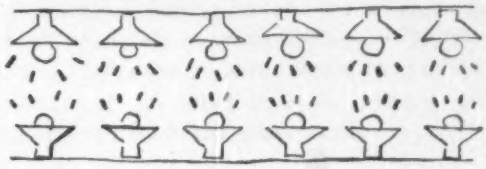
High-frequency Dielectric Heating

This form of electro-heat can be used only on non-conducting materials such as wood, plastic and rubber. The material is placed between two electrodes to which a high voltage is applied at a high frequency. This has the effect of generating heat inside the material rapidly and uniformly throughout its entire thickness.



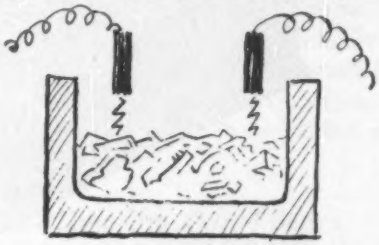
Infra-red Heating

This method employs pure radiant heat. The bulk of the radiation takes place in the infra-red portion of the radiation frequency spectrum. The heaters may take the form of reflector lamps or sheathed wire elements. The method is extremely flexible and has many uses, including paint drying and pre-heating plastics.



Arc Melting

This form of heating is chiefly used for melting steel. The diagram illustrates one method of operation.



Very large charges can be melted; melting units of 200 tons capacity are now in operation.

The Application of Electro-Heat

All these methods of electric heating can be applied in almost an infinite variety of ways. Some of these ways will be dealt with in subsequent sheets.

For further information, get in touch with your Electricity Board or write direct to the Electrical Development Association, 2 Savoy Hill, London, W.C.2.

Excellent reference books on electricity and productivity (8/6 each or 9/- post free) are available—"Induction and Dielectric Heating" is an example; "Resistance Heating" is another.

E.D.A. also have available on free loan a series of films on the industrial use of electricity. Ask for a catalogue.

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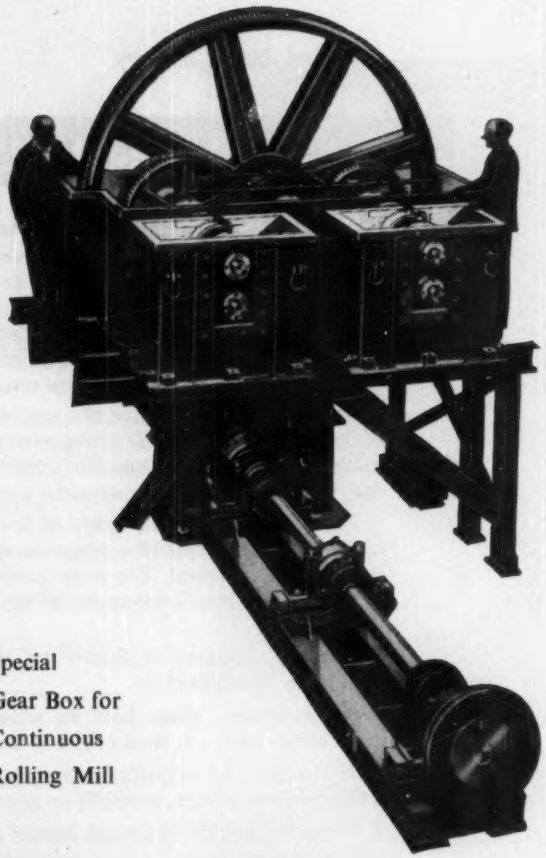

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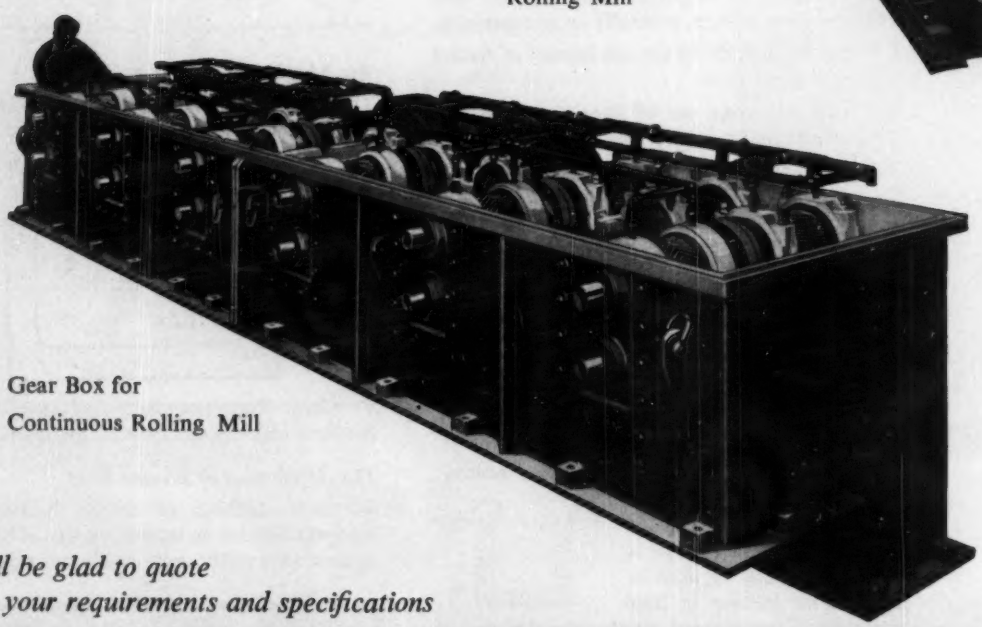

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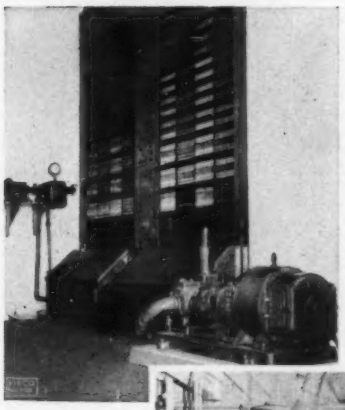
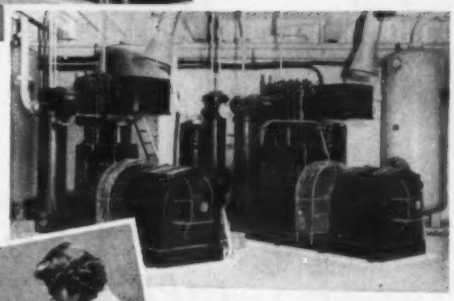
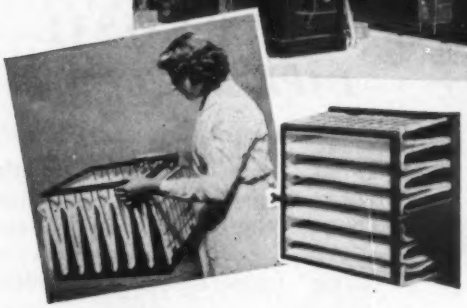
VISCO
AIR FILTRATION



Left: One of five "Reciprojet" Self-Cleaning Air Filters, total capacity 118,000 c.f.m. at the Albion Works of The Briton Ferry Steel Co. Ltd.





Above: Five "Broomwade" Air Compressors fitted with "Visco" Static Cell type Air Filters, at the Luton Factory of Vauxhall Motors Ltd.

Bottom: "Visco" "M.V." Dry Cell Air Filter. The photo on the left shows the ease with which the filter medium can be fitted. That on the right shows a complete unit.

For nearly 40 years we have been designing and manufacturing Air Filters for use in the ventilating and air conditioning systems of buildings, ships, rail coaches and the ventilation of electrical equipment in steelworks, power stations, collieries, etc. "Visco" Air Filters are divided into three main groups.

"Static" Cell oil-wetted Filters

These are robust units of all-metal construction and of standard dimensions and can be built up in panels of any air capacity desired. "Visco" Static Air Filters are used the world over in the ventilation of buildings and on diesels, air compressors and diesel/elec. locos.

"Dry" Cell Filters

This type is designed for use in the air conditioning of operating theatres, research laboratories, photo processing rooms and other places where an atmosphere entirely free from dust is essential. In some installations "Visco" Dry Cell Filters are used in conjunction with our Static Cell type. Both are of the same size.

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The "Visco" Patent "Reciprojet" Self-Cleaning Air Filter has proved most successful for the ventilating of elec. machinery in steelworks, collieries and other places where the air is heavily laden with dust. Its salient features are :- two-stage filtration, auto time switch control, cleaning of cells by reciprocating oil jets and positive prevention of oil carryover.

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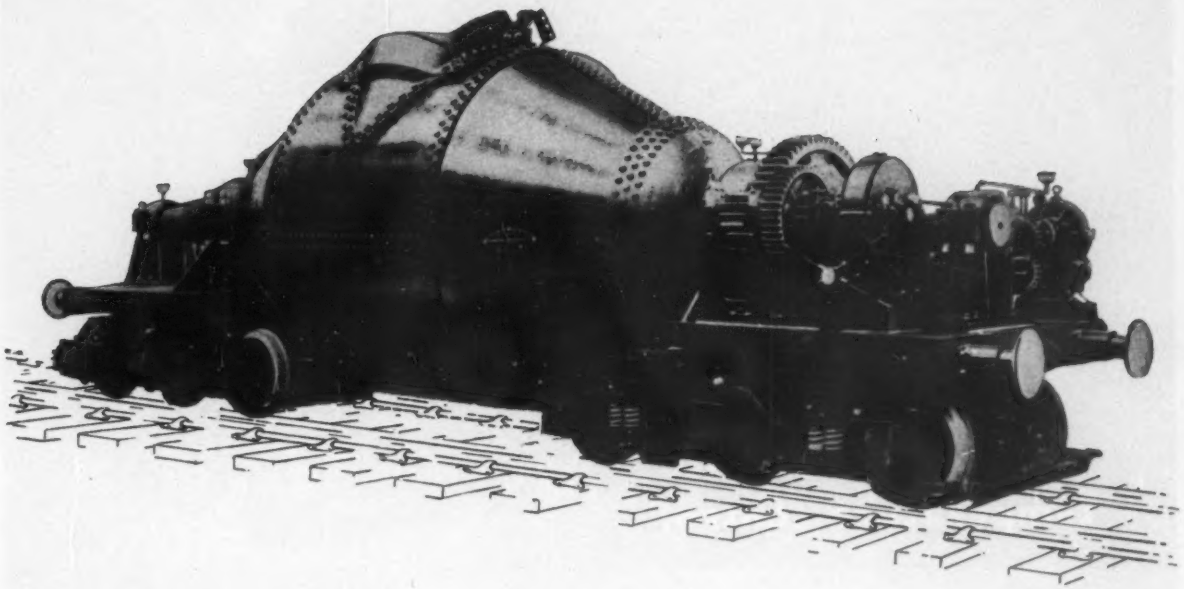

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Photograph of a quick immersion thermocouple in use in an open hearth furnace, shown by courtesy of Appleby-Frodingham Steel Co. Ltd.

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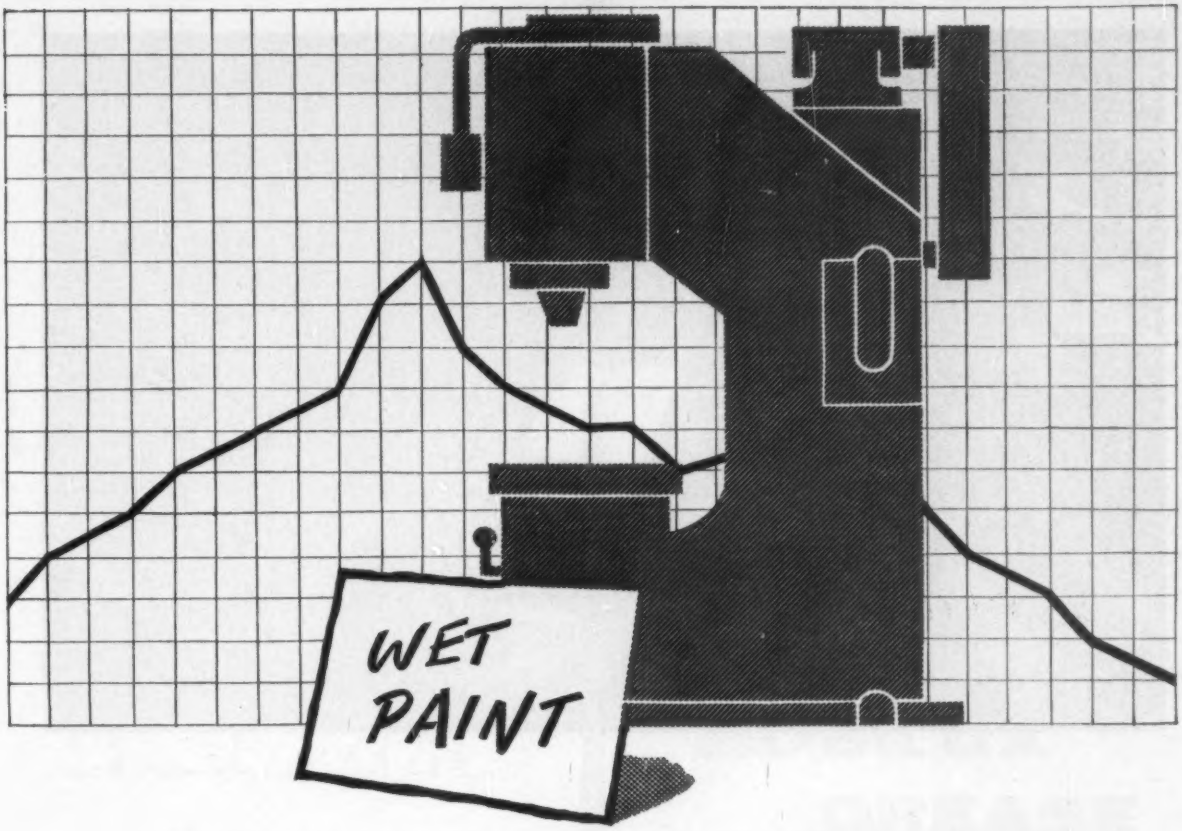

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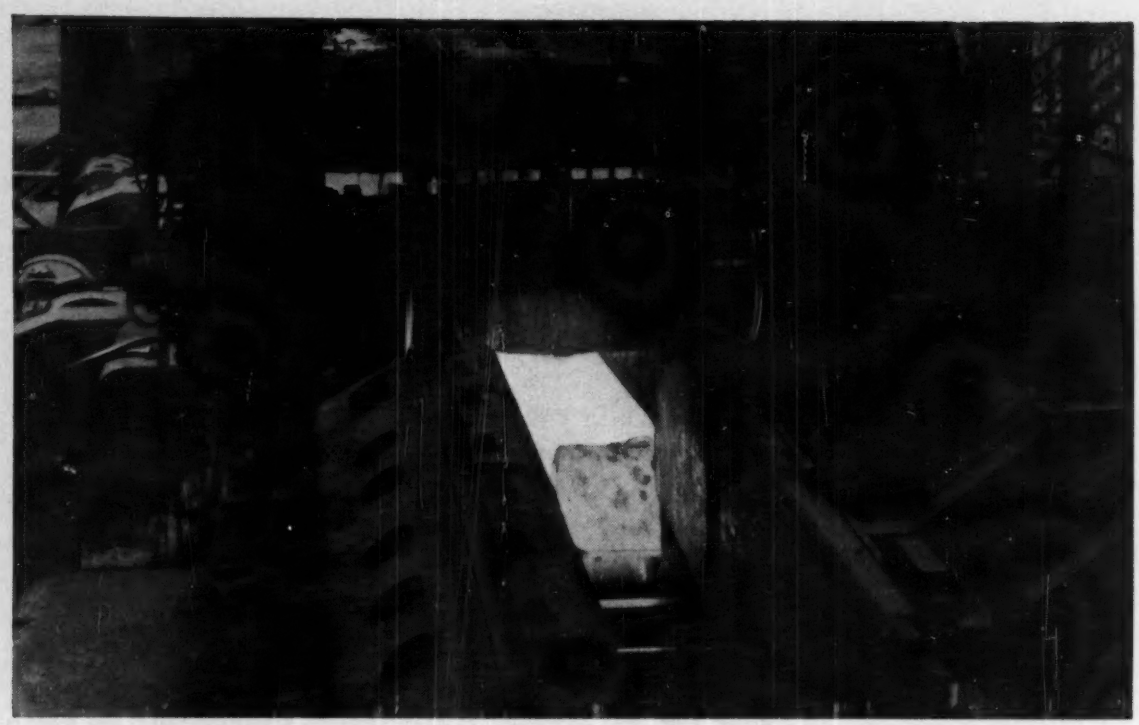

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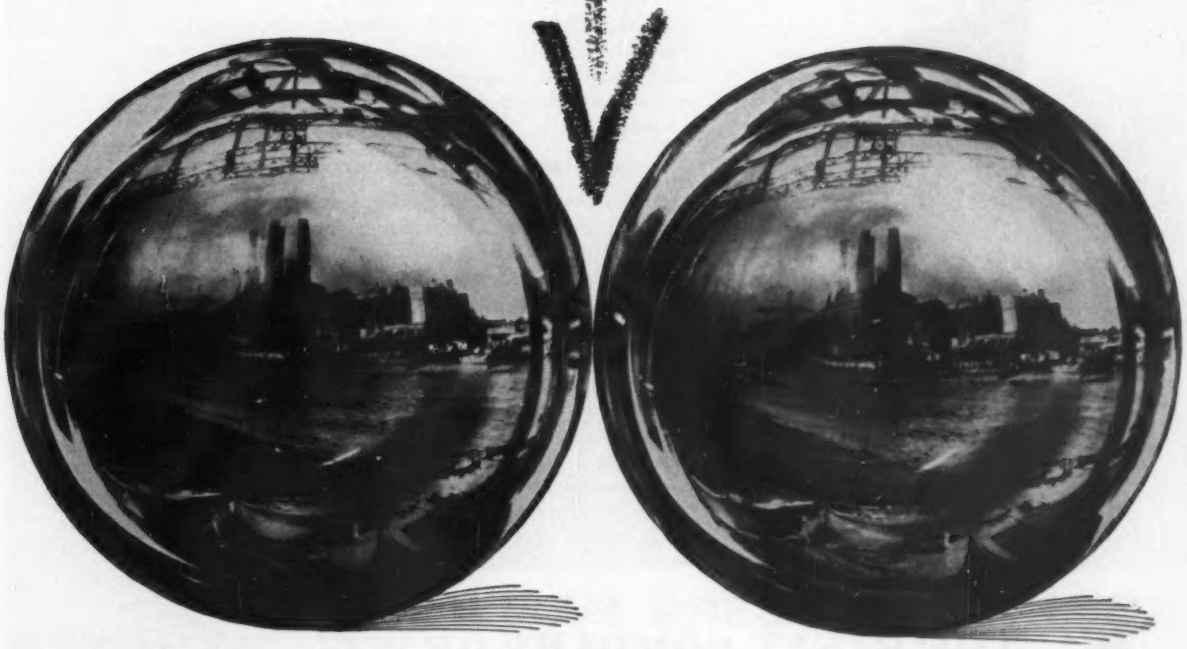

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It has always been the policy of The Steel Company of Wales to pay particular attention to customers' specific problems, and to ensure that its products are "tailor-made" to individual requirements. If you have an industrial problem which sheet steel might help to solve, it will be worth your while to write to us or telephone Port Talbot 3161. We believe we can help.

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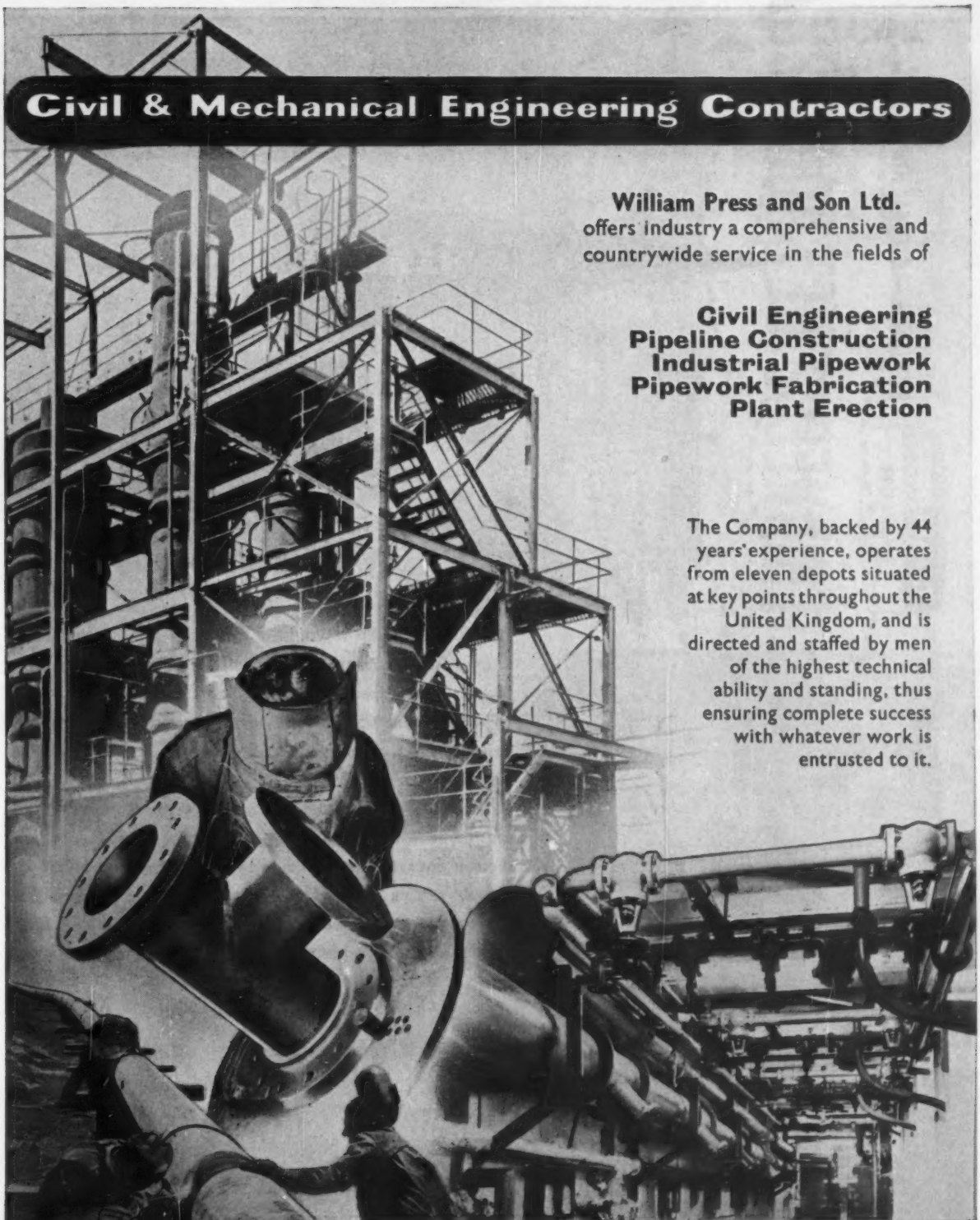

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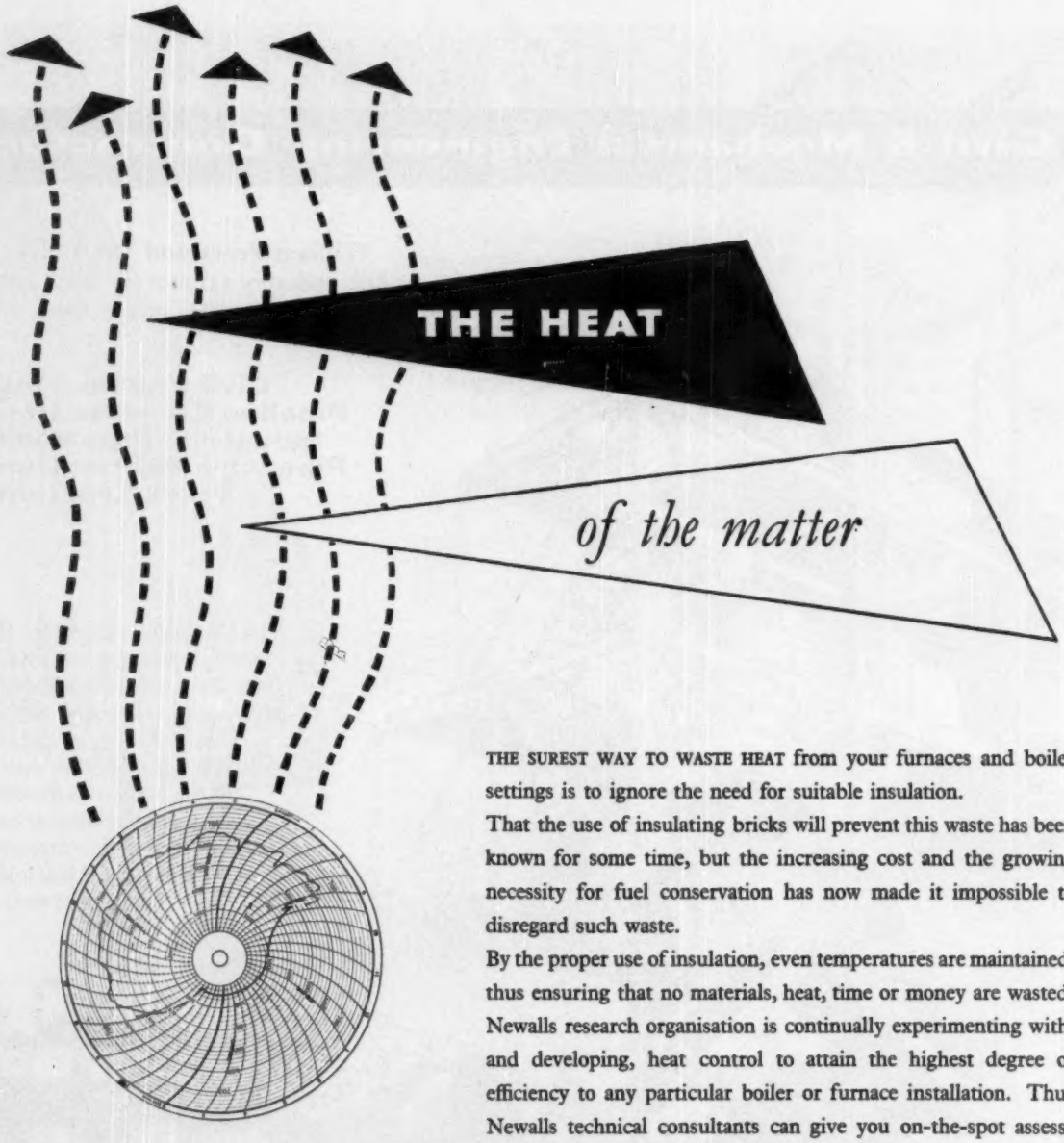

THE SUREST WAY TO WASTE HEAT from your furnaces and boiler settings is to ignore the need for suitable insulation.

That the use of insulating bricks will prevent this waste has been known for some time, but the increasing cost and the growing necessity for fuel conservation has now made it impossible to disregard such waste.

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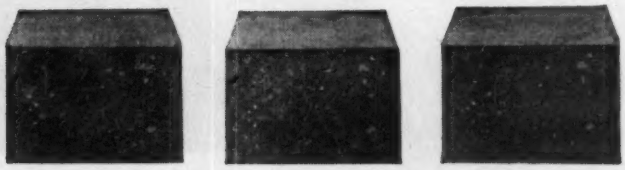

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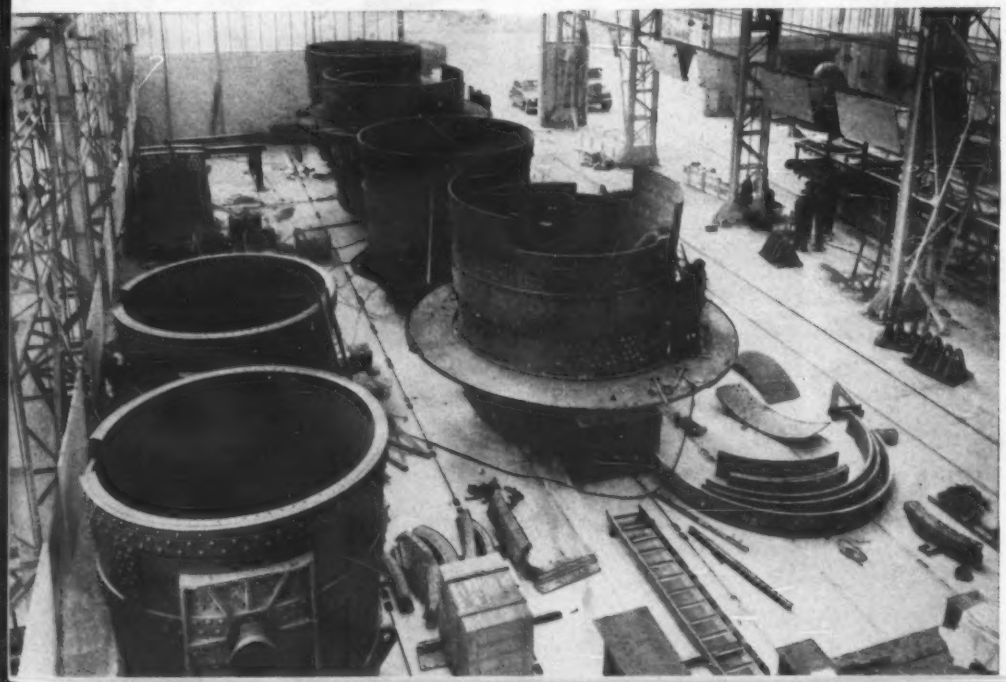

RIGHT *One of many steel casting ladles supplied to John Summers and Sons Limited.*

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Hunnex is made in standard squares, open hearth and electric furnace roof blocks and special shapes for the Glass and Carbonising Industries.

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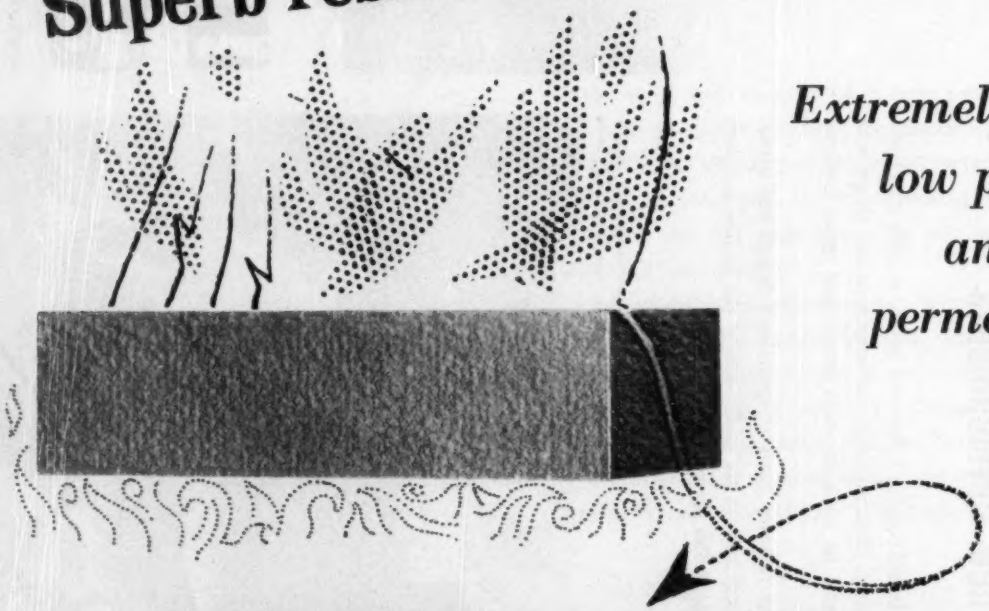

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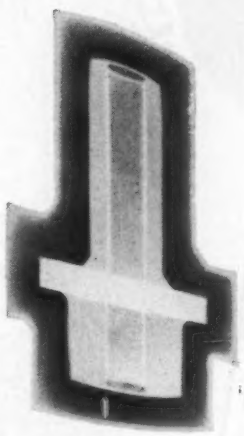

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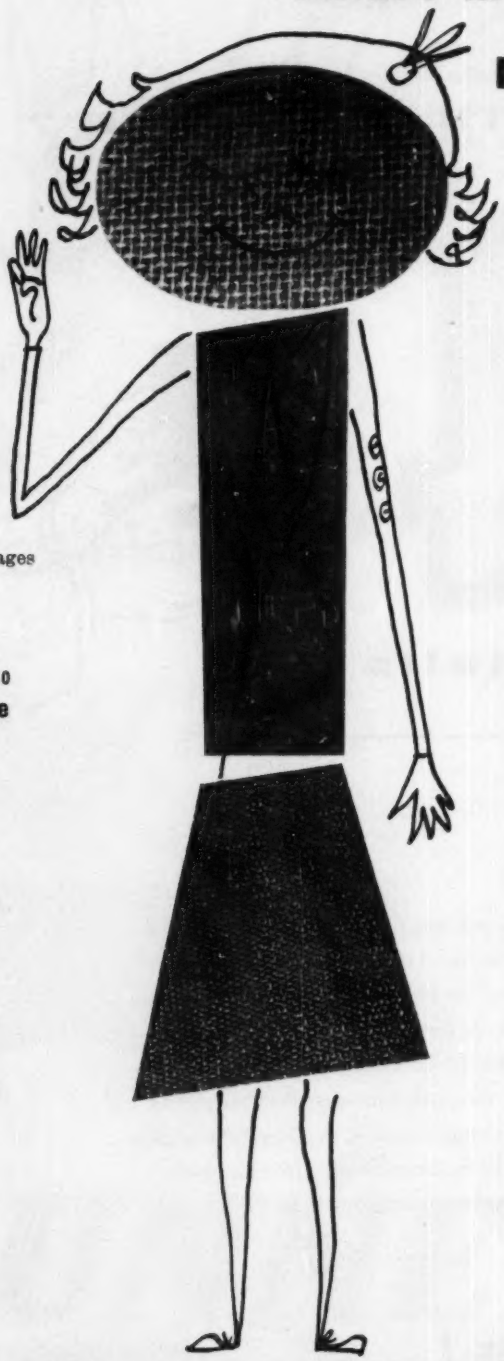

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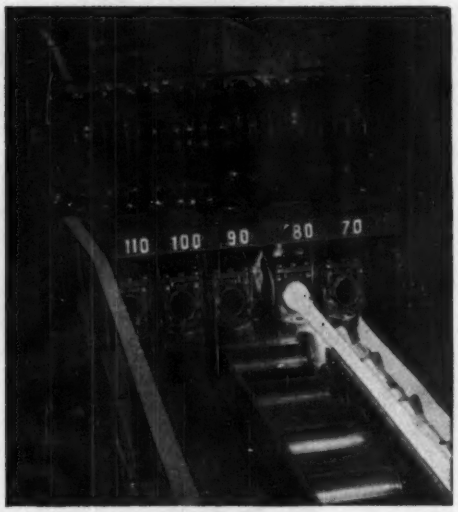
8 good reasons why you should use these ROLLER GUIDES in your bar or wire rod mill



Nos. 125 and 155 roller guides fitted in a 27 1/2 in. (700 mm.) diameter, three-high stand for rolling 2 3/4, 3 3/16, 3 9/16, 4, 4 3/8 and 4 7/8 in. (70, 80, 90, 100, 110 and 124 mm.).

1 The entry friction guides are rigidly clamped into the guide box. They are intended to lead the oval to the groove in the roller, to protect the rollers against overloading, and to straighten bent rods.
The groove in the entry guides is made about 1/8" wider than the oval.

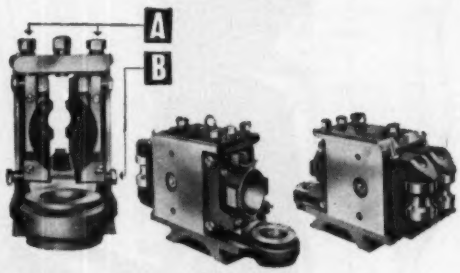
2 The rollers are mounted on leaf springs, which can be adjusted by means of the screws "B", and hold the oval steadily even if there should be some slight variation in the thickness of the oval.

3 If the thickness of the oval should vary, the springs of both rollers yield an equal amount, so that the oval remains in the centre of the groove.

4 To tilt the oval, the screws "A" are adjusted. One roller will then be lifted as much as the other is lowered.

5 As the oval is held very rigidly, the leader oval for wire rod may be thick, unless other circumstances call for a thinner oval.

6 The roller guide assemblies are narrow, and all screws that require adjustment during rolling are accessible from the front of the mill. No lateral space outside the guide box is therefore required for gaining access to the set screws, and the guide box assembly can thus be located close to the mill housing.



7 As the groove in the entry guides is wide and nothing but rolling friction acts on the oval, the latter is very easily introduced into the pass, and there is practically no risk of scratching the bar.

8 The roller guide may be used not only for leader and drawing oval passes but also for edging flats as well as for rolling squares, hexagons, octagons and various other sections, for instance in reduction passes for hexagon. It is suitable for all types of rolling mill. An important advantage is that the roller guide enables repeaters to be used in rolling plump leader-ovals.

PS *Practically speaking, every bar and wire rod mill in Scandinavia uses roller guides of our design.*

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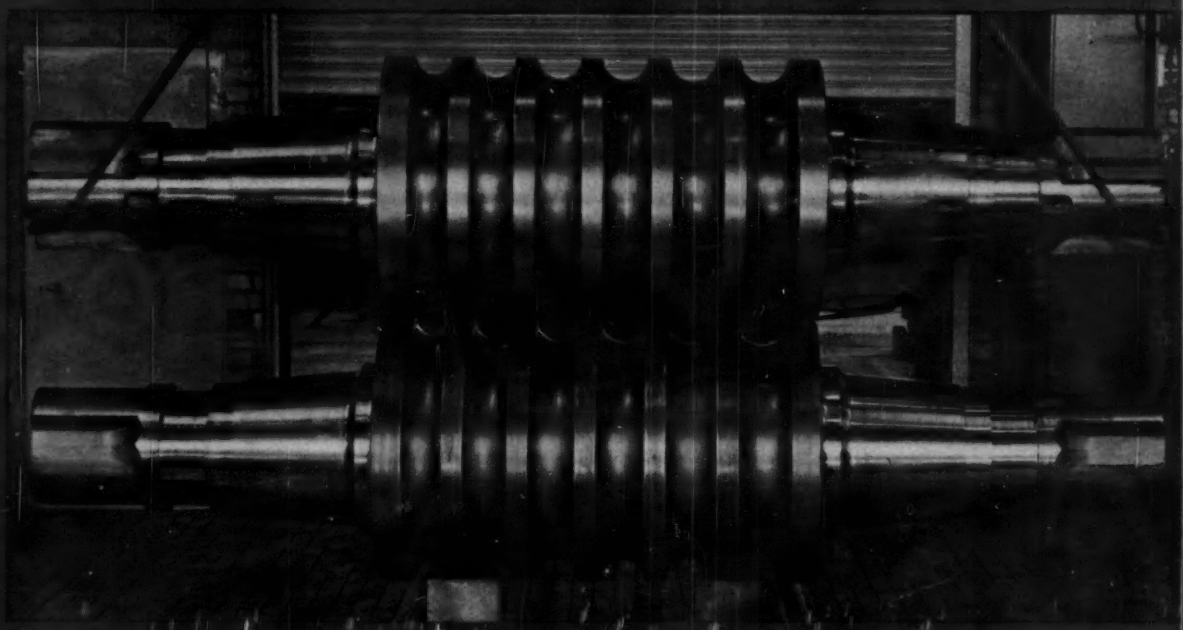
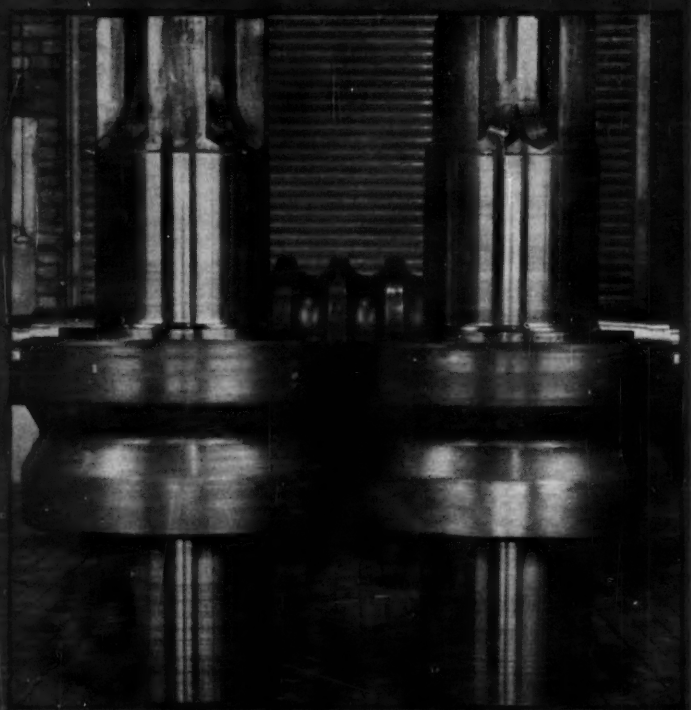


A pair of the Special 9"/8½" diameter Tandem Rolls for Finishing Large Rounds 4½" to 6" dia. for No. 6 Stand Rod Mill 31" dia. x 12" long barrel.



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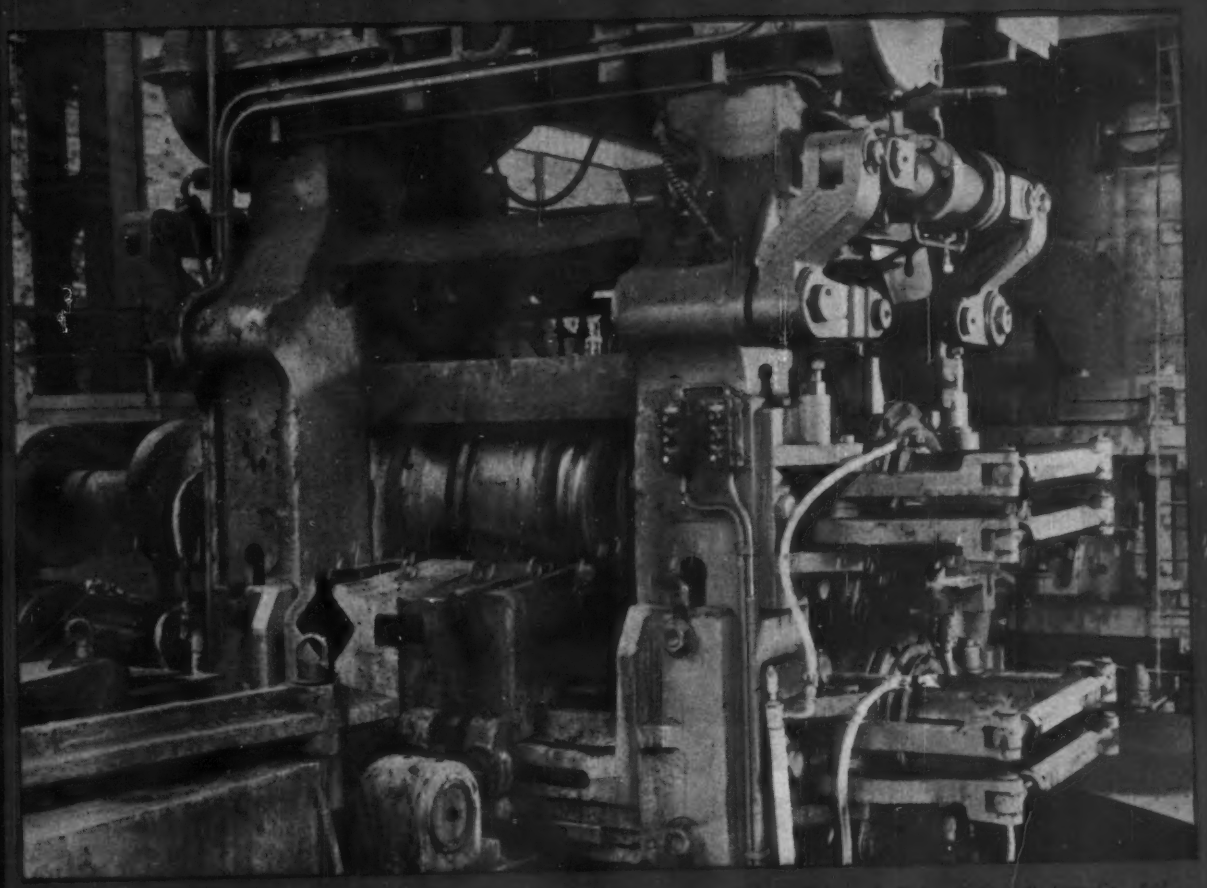

No. 6 Stand Forming for 4", 4½" and 5" Billets, also Breaking for 7", 8", 9" Billets and Slabs at Consett Iron Co., Ltd., showing our Special S.S.X. Mooralloy Quality Rolls. Roll size 32" dia. x 54" Long Barrel.

Photographed by the courtesy of The Consett Iron Co. Ltd.

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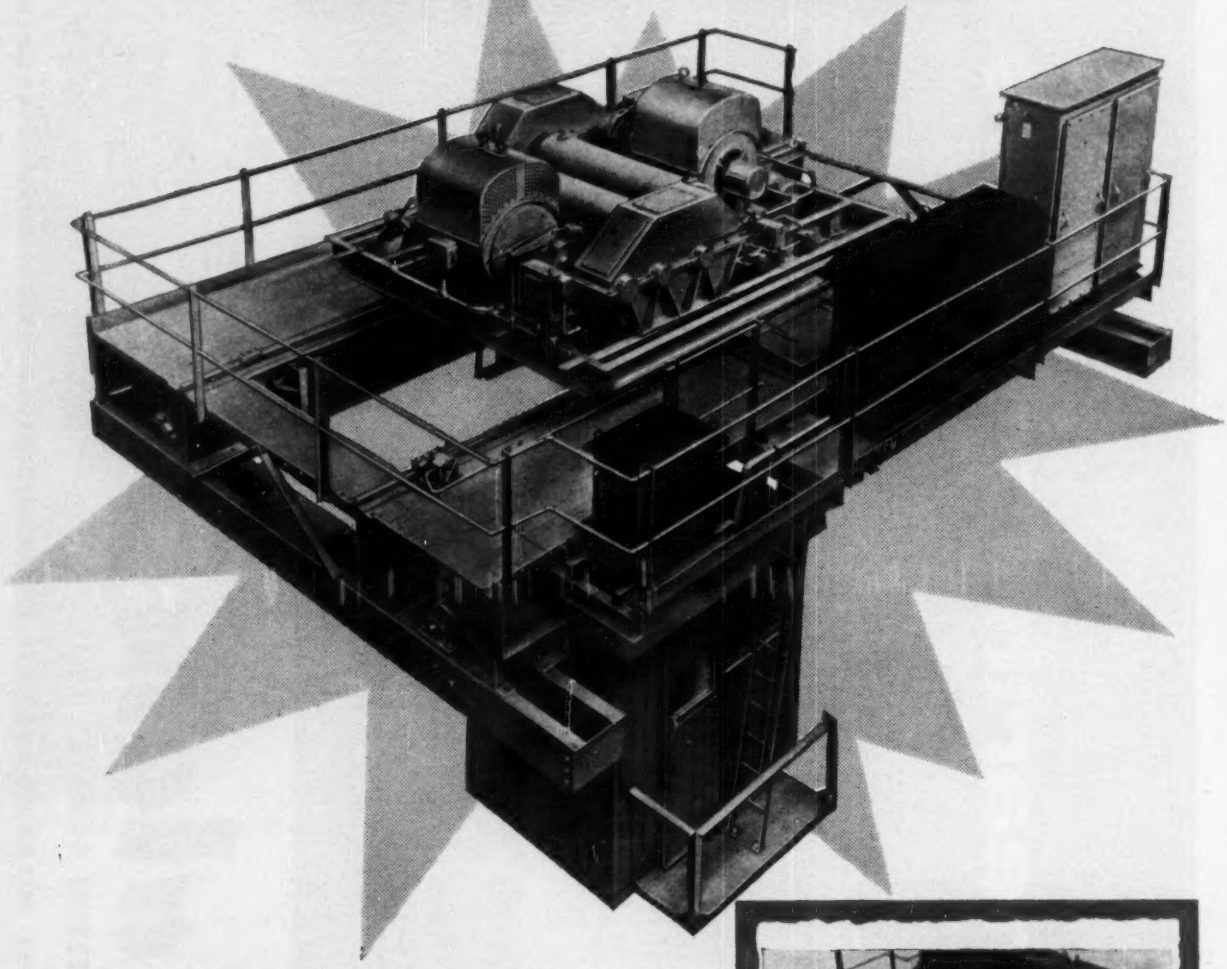
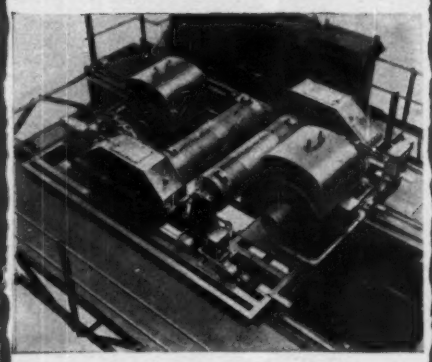
OVERHEAD
GRABBING CRANE



This crane is one of two specially designed for heavy duty grabbing service in one of the new super power stations. The Cranes are operating in the open, and special attention has been given to weather-proofing and positioning of the units for easy access.



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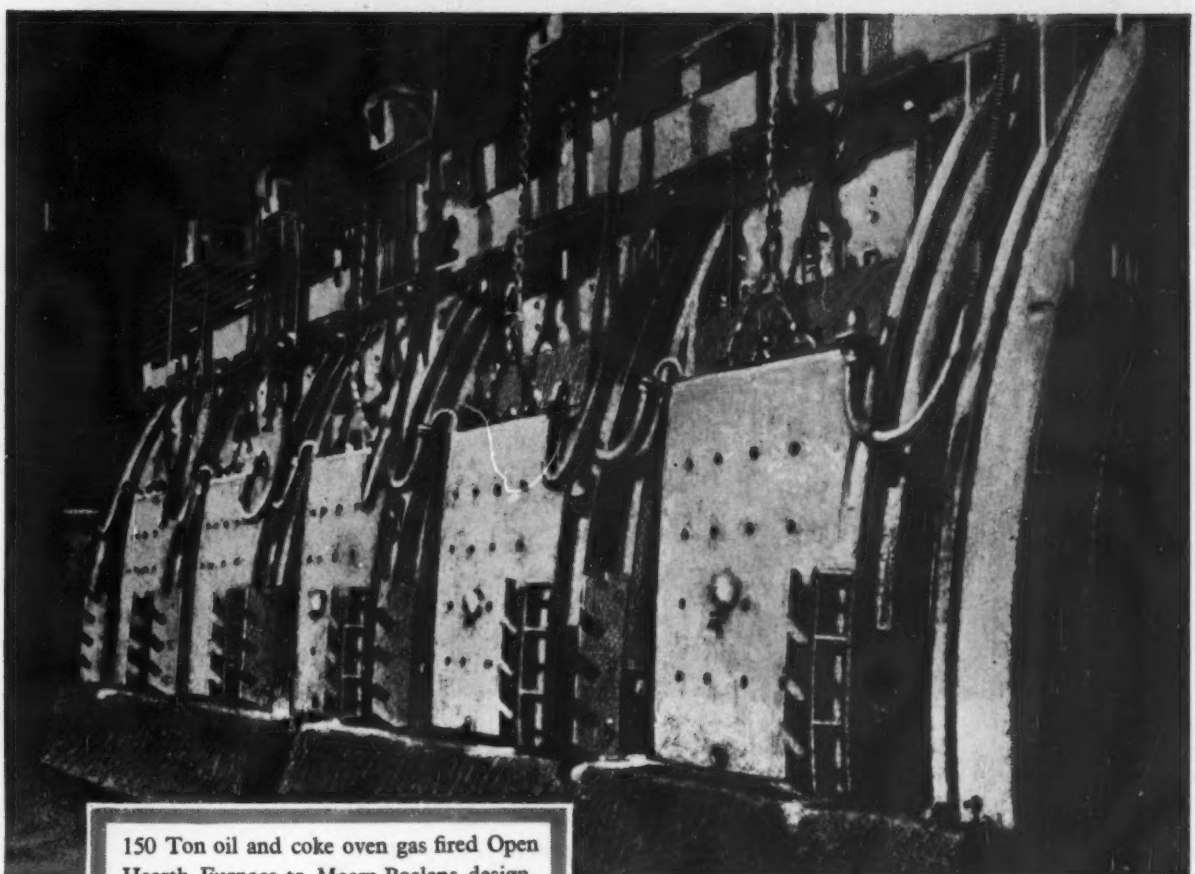

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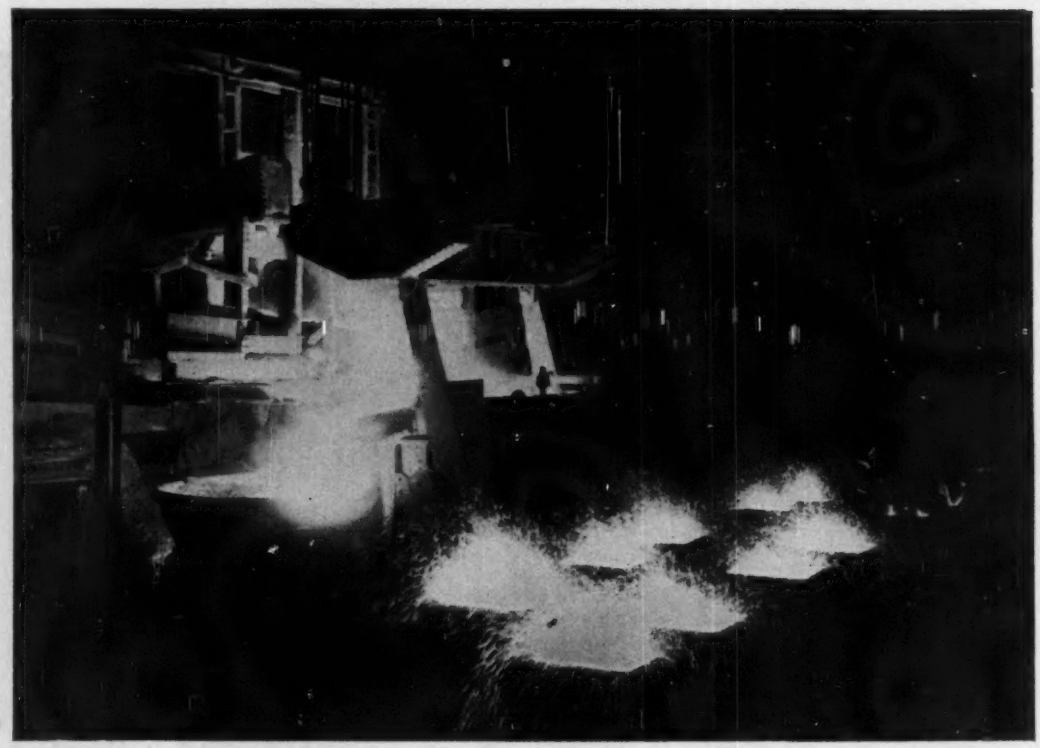

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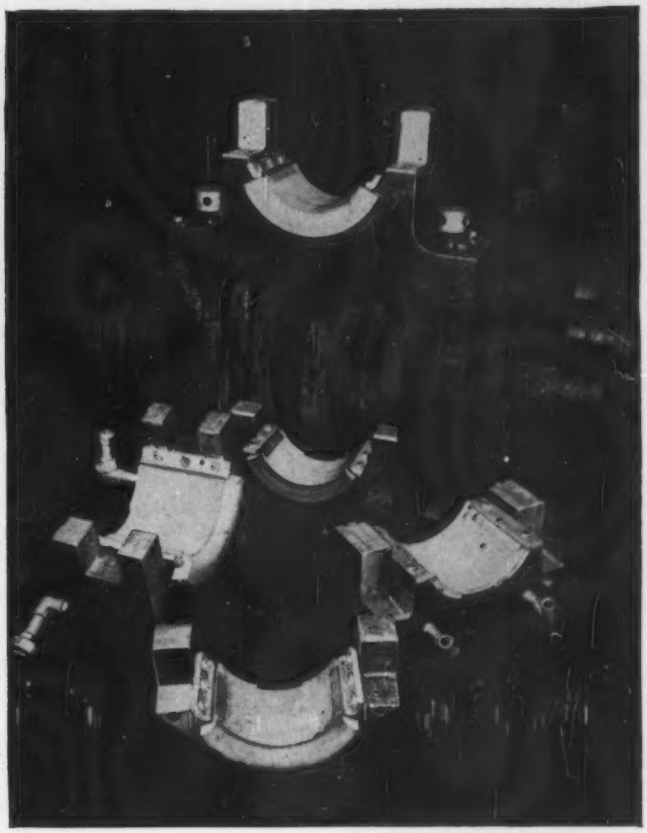

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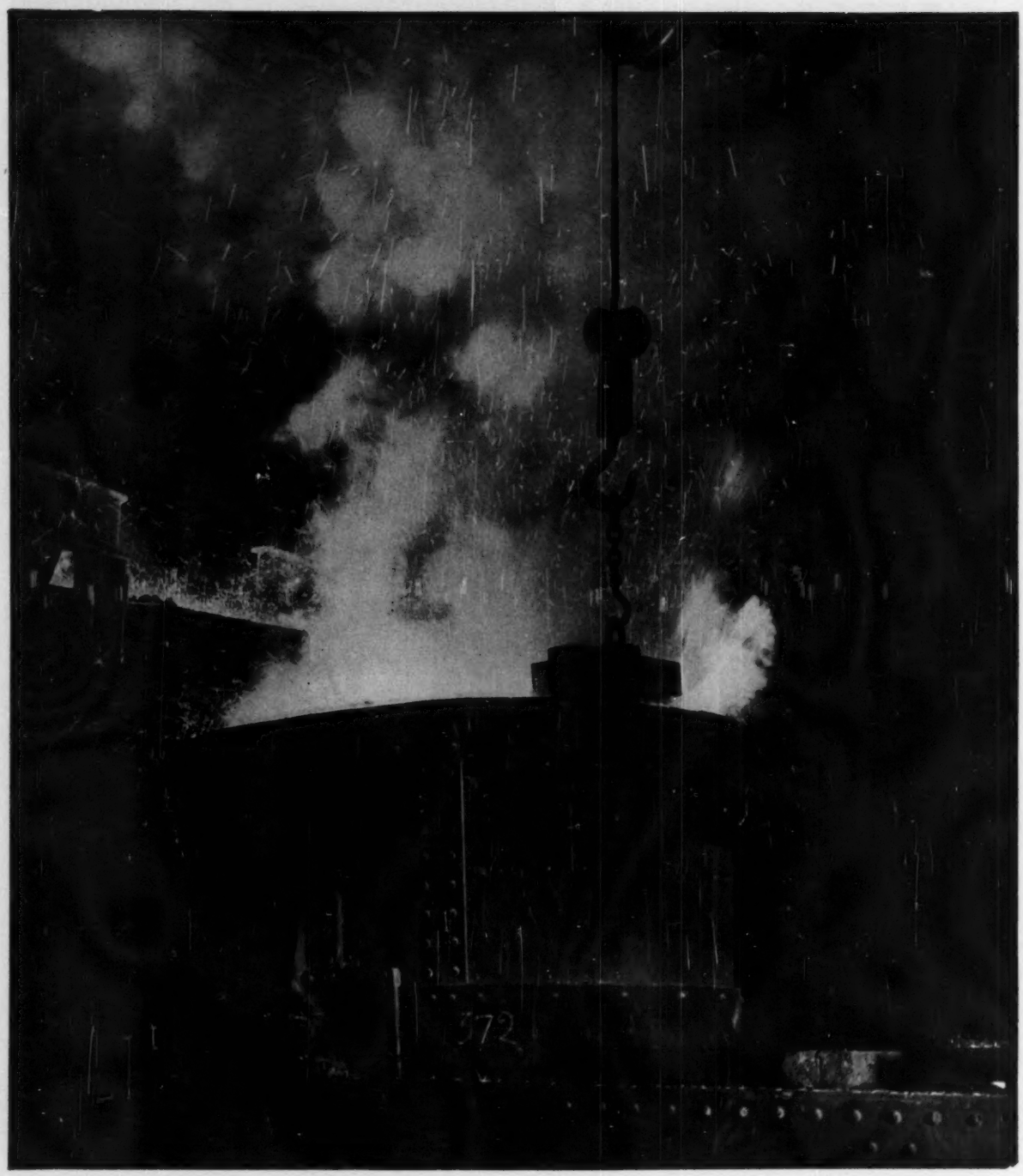

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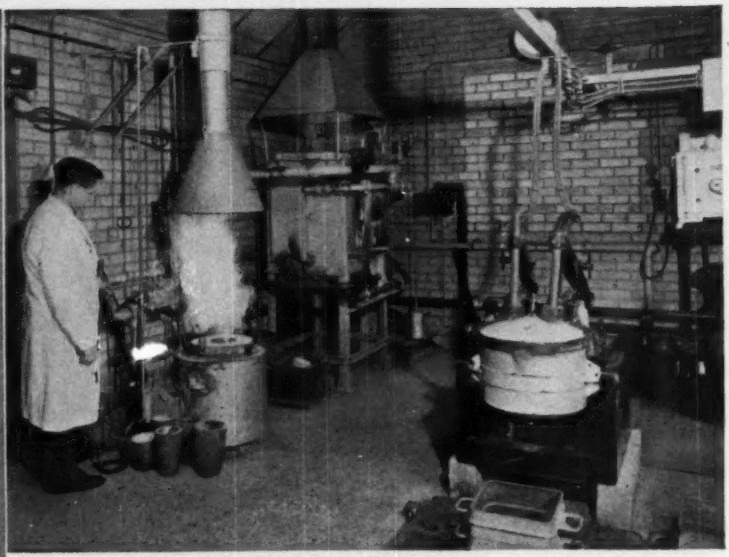

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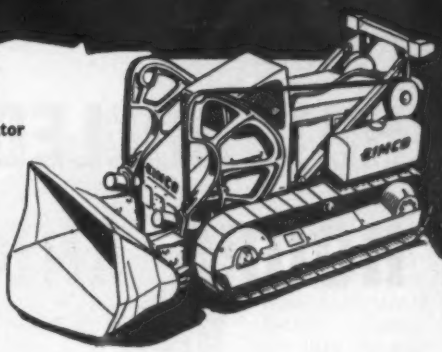

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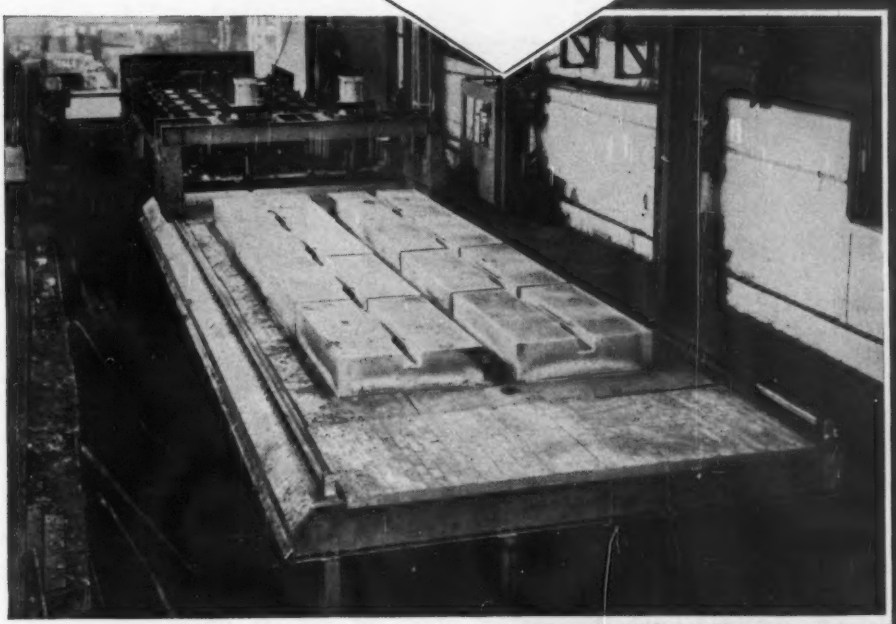

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Electric soaking pit, designed and built by Stein and Atkinson Ltd., for the Appleby—Frodingham Steel Company (Branch of the United Steel Companies Ltd.), Scunthorpe

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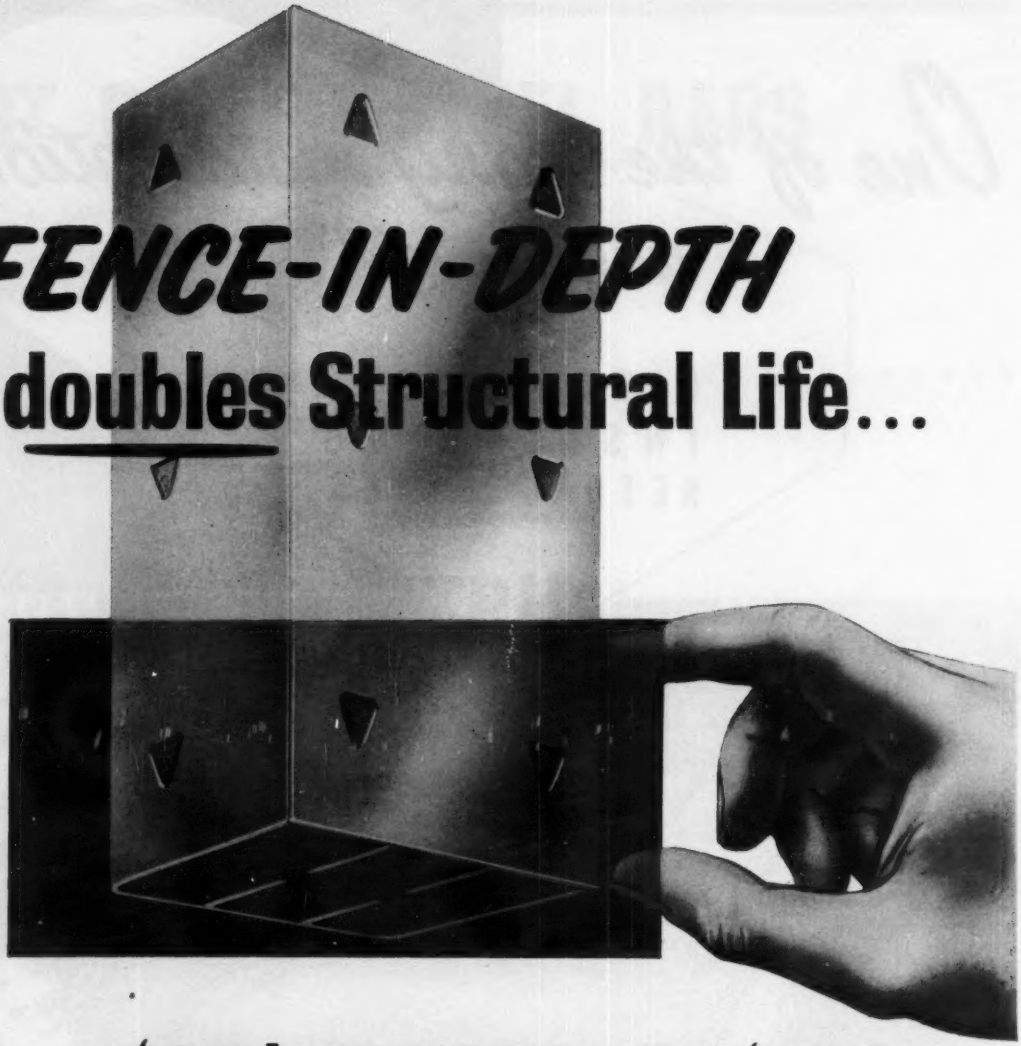

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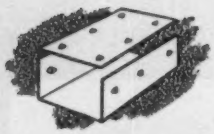

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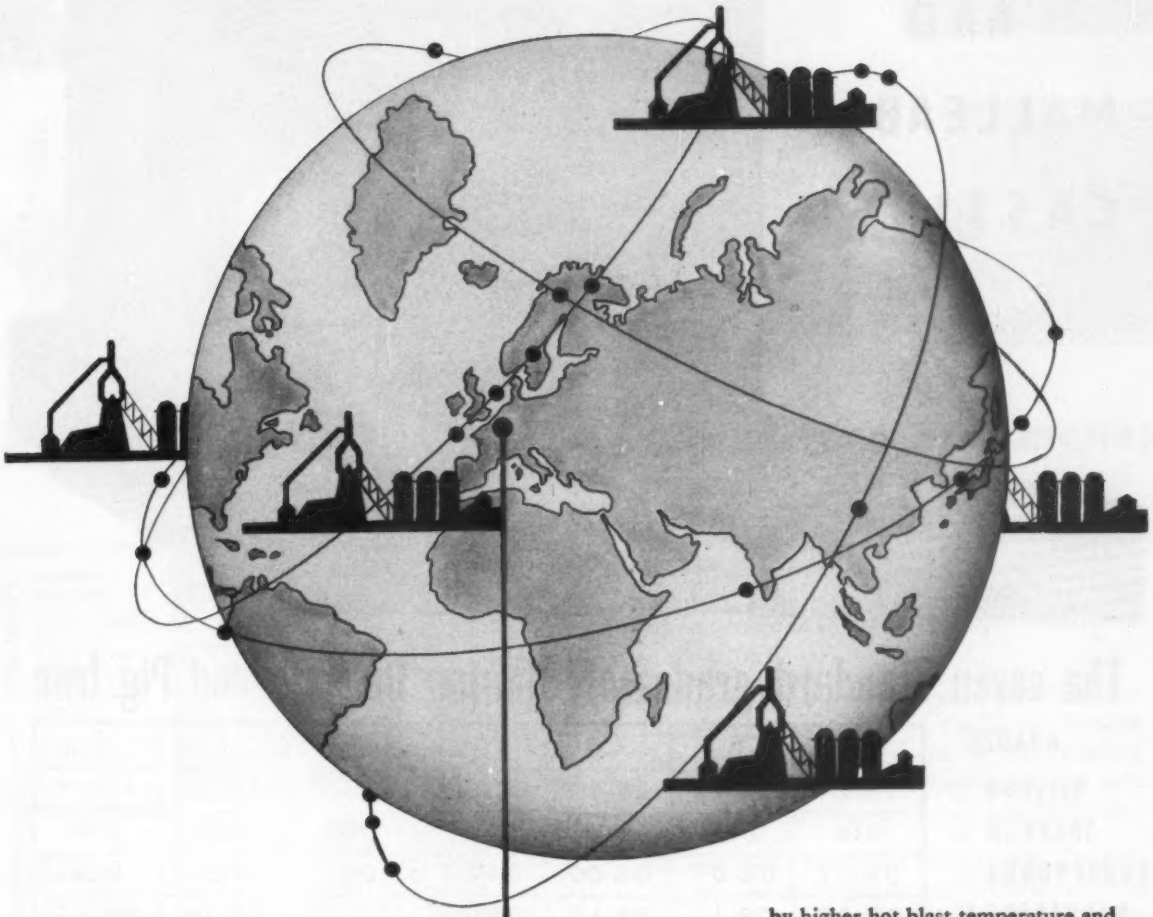

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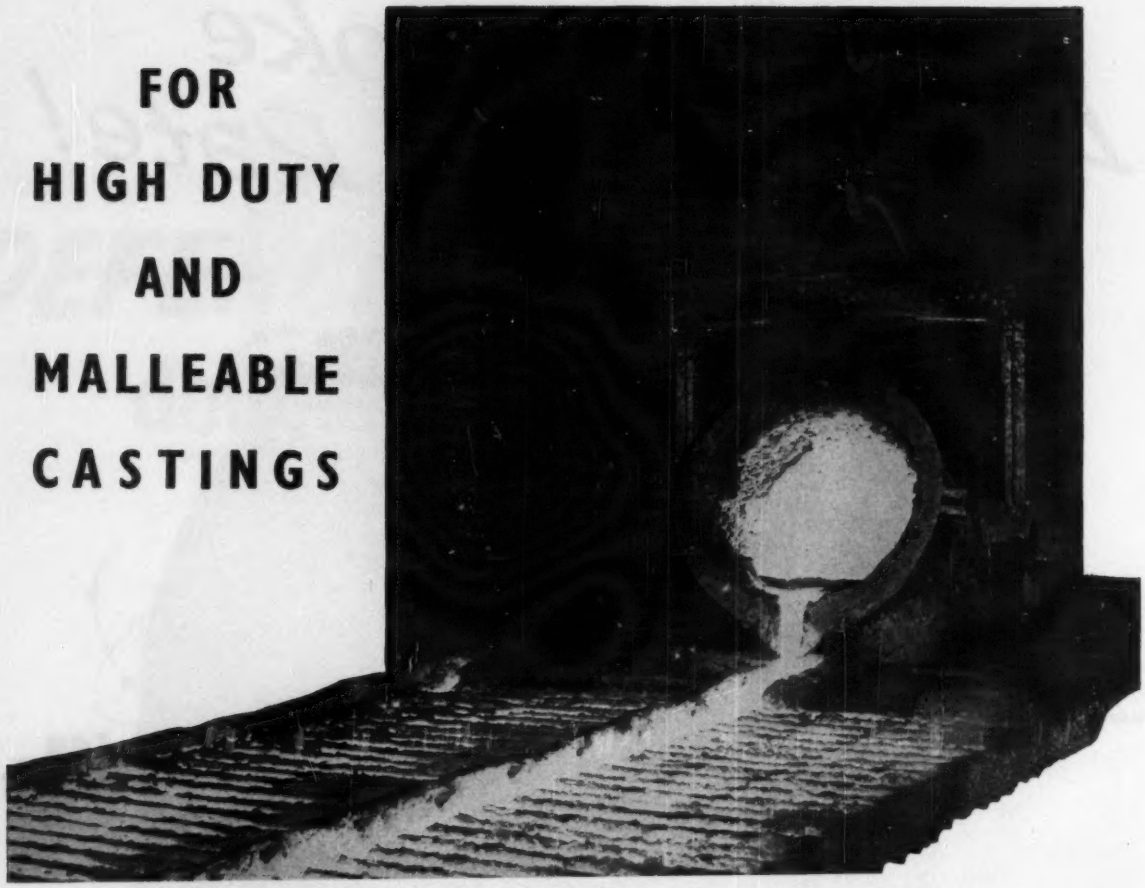

The seven standard grades of Stanton Dale Refined Pig Iron

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SILICON %	1·9–2·1	1·4–1·6	0·9–1·1	2·25–2·5	0·9–1·1	1·4–1·6	1·9–2·1
SULPHUR %	0·06	0·07	0·08	0·05	0·08	0·07	0·06
PHOSPHORUS %	0·6–0·7	0·6–0·7	0·6–0·7	0·4–0·5	0·25	0·25	0·25
MANGANESE %	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5	0·8–1·5
TOTAL CARBON %	2·7–3·0				2·8–3·1		

In addition to the above other grades of Dale Refined Iron can be supplied to meet more exacting demands which call for alloy additions and special processing.

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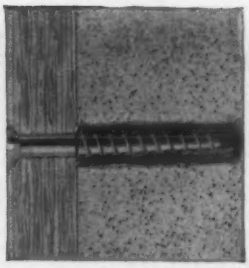

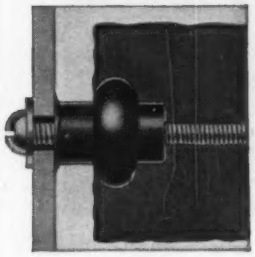

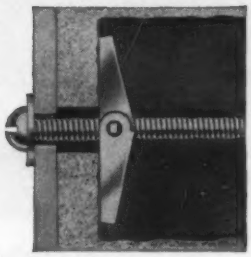

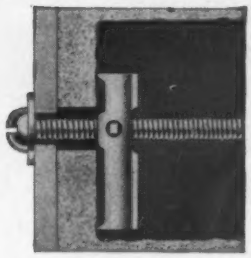

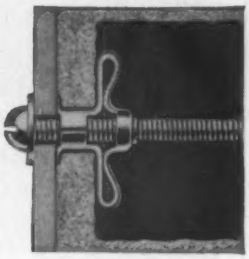

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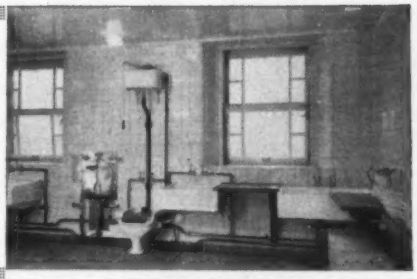

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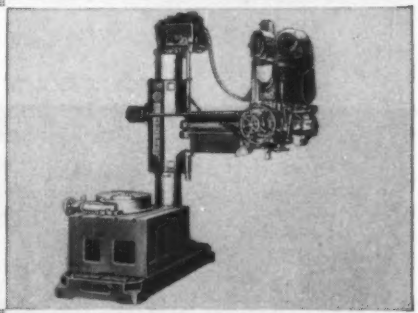

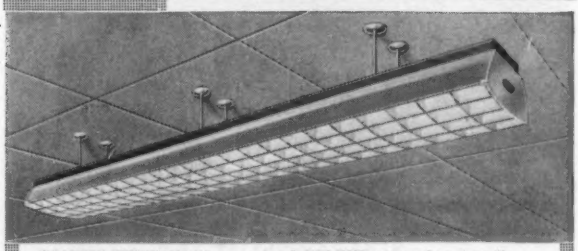

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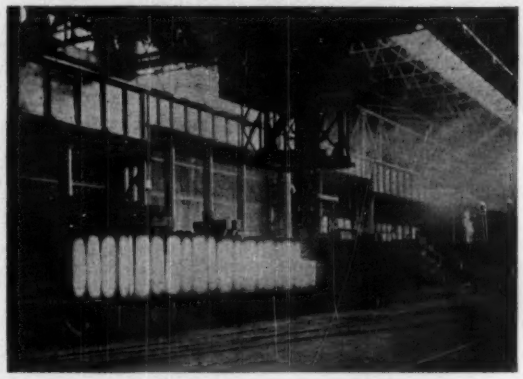

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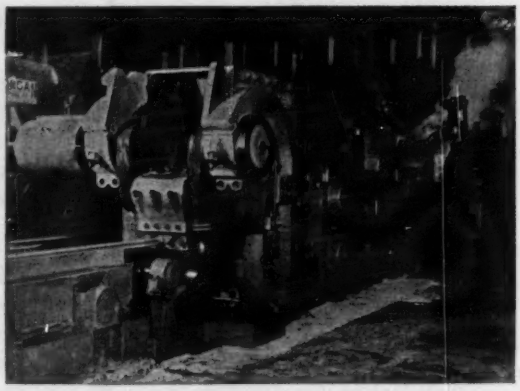

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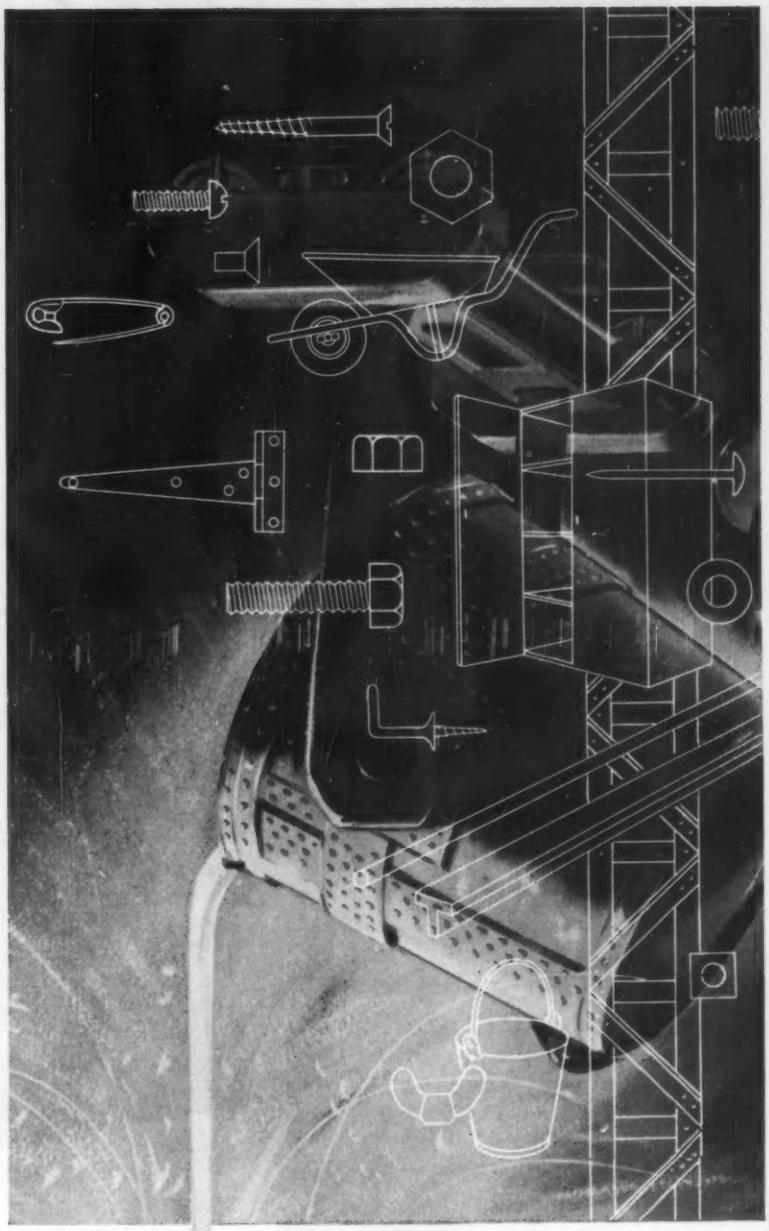

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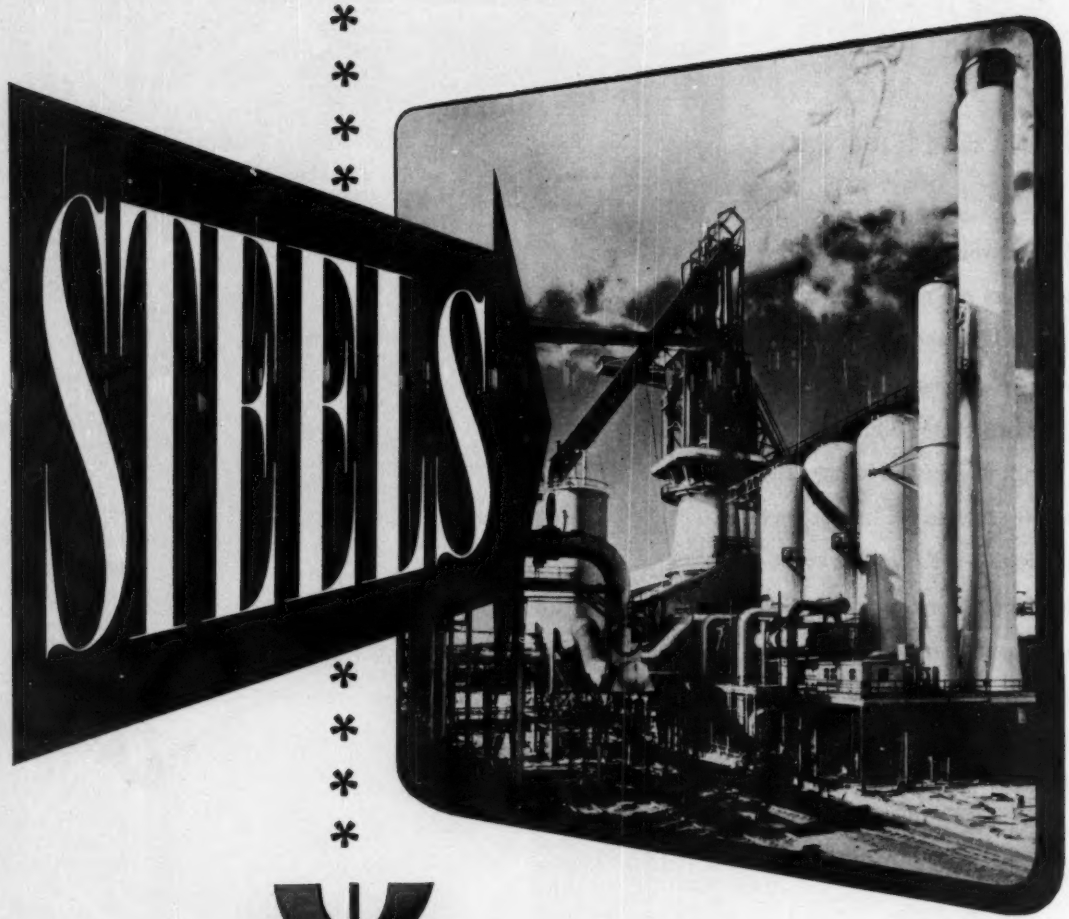

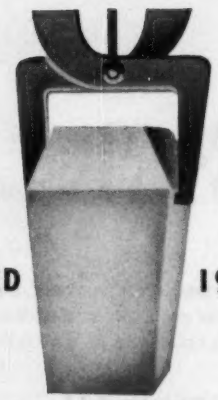

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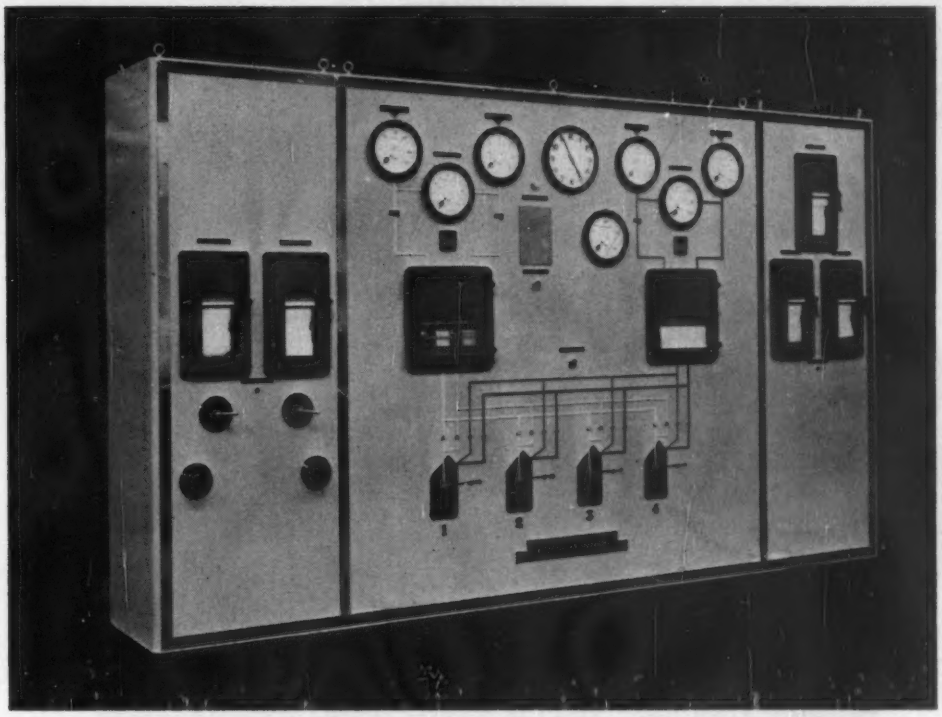

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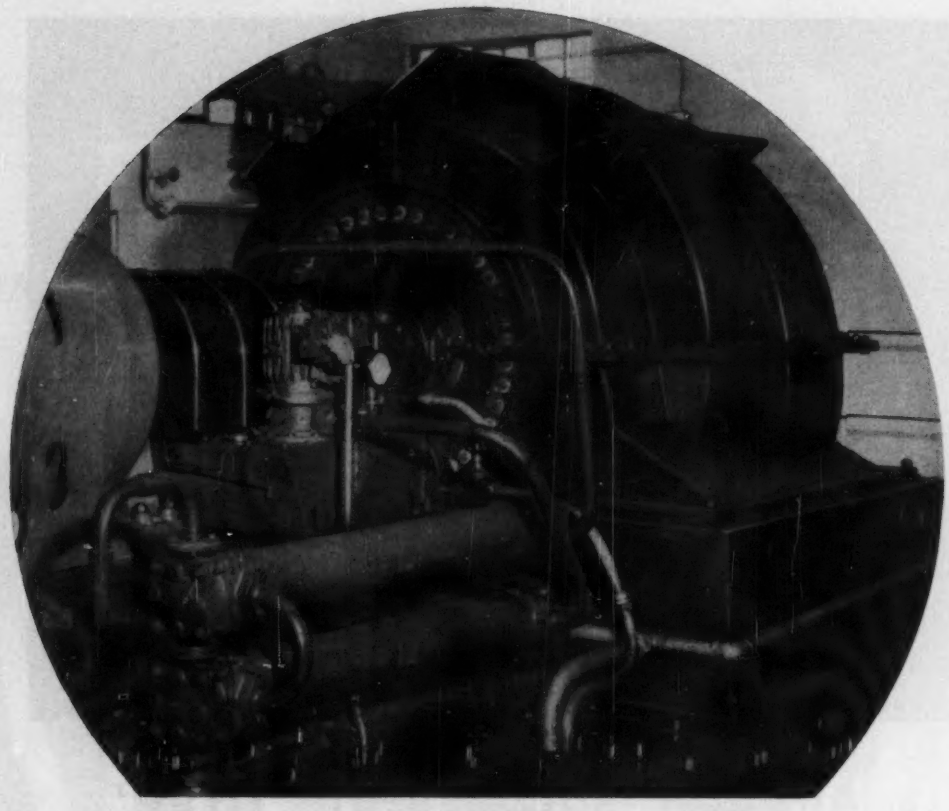

TWO-STAGE GAS BOOSTER

The illustration reproduced by the courtesy of the West Midlands Gas Board shows the Howden two-stage Booster with variable speed drive.

This installation is at Washwood Heath Gas Holder Station, and is designed to handle an inlet volume of 2,250,000 cubic feet per hour of Town's gas at a differential pressure of 80" wg. The total adiabatic efficiency of the Booster is 82%

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- Uniformly graded coke of constant fineness so essential for making first class sinter.
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- Reliable operation with lowest overall maintenance costs.
- Proved by continuous service since 1938 in seven leading British Steelworks.

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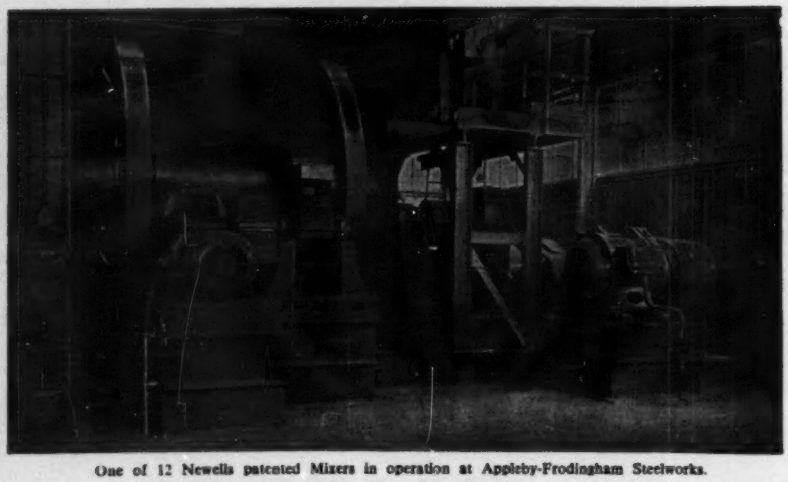
- We have supplied 37 Feed Tables for different sinter plants at Messrs. Appleby-Frodingham Steel Company.



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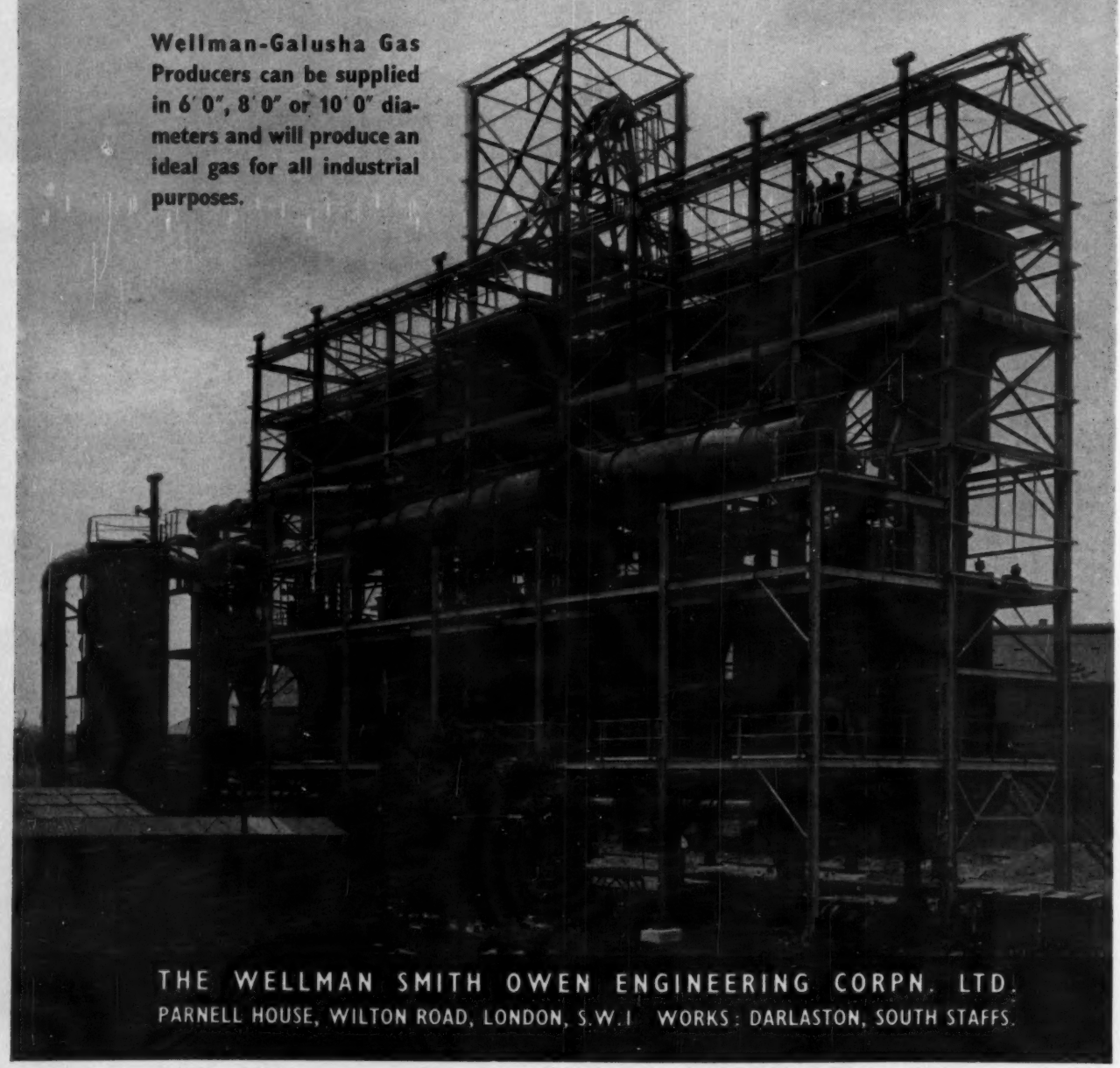

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HAROLD MOORE, C.B.E., D.Sc.

HAROLD MOORE was born and brought up in Middlesbrough, and it was there that, in 1894, he became a pupil as a metallurgical chemist of J. E. Stead. He obtained the B.Sc. degree of London University in 1898, and then spent two years as chemist in a small Northamptonshire ironworks. In 1900, on Stead's recommendation, he was appointed Research Metallurgist at the Parkhead steelworks of William Beardmore and Co., Ltd. Four years later Dr. Moore joined the Research Department, Woolwich, as Chief Metallurgist.

The First World War brought innumerable problems, and by 1918 his establishment had become one of the country's leading metallurgical laboratories. In 1920 he was appointed Director of Metallurgical Research. In the years between the wars, Dr. Moore and his staff were engaged on work for both the Services and for Industry, and many papers were published in metallurgical and engineering journals. During his period at Woolwich, Dr. Moore served on the Ordnance Board, the Aeronautical Research Committee, and many other Service, B.S.I. and industrial committees, including three of the Joint Research Committees set up by the British Iron and Steel Federation and The Iron and Steel Institute; he was the first Chairman of the Steel Castings Research Committee.

In 1932 Dr. Moore succeeded Dr. R. S. Hutton as Director of the British Non-ferrous Metals Research Association. After his retirement in 1944 he was Chairman of an Admiralty metallurgical research committee, served on the Inter-services Metallurgical Research Council, and acted as a consultant to B.I.S.R.A. He took an active part in the formation of The Institution of Metallurgists in 1945, and was its first President; in 1946-1947 he spent some months in India, advising the Tata Iron and Steel Co., Ltd., on the organization of scientific control and research at their Jamshedpur works.

A member of The Institute of Metals since 1920, Dr. Moore served on the Council for ten years before being elected President for 1934-1936. In 1937 he was awarded the Institute's Platinum Medal and elected a Fellow of the Institute. He is a Founder Fellow of The Institute of Physics, and has served on the Councils of the Royal Institute of Chemistry and the Faraday Society. He became Honorary Treasurer of the Society of Instrument Technology in 1949, and also undertook the editing of the Society's quarterly Transactions; on giving up this work in 1956, he was elected an Honorary Life Member.

Dr. Moore joined The Iron and Steel Institute in 1901, and read his first metallurgical paper at one of the Institute's meetings in 1910. The Council elected him an Honorary Member of the Institute in September, 1958.

Harold Moore, C.B.E., D.Sc.

Honorary Member

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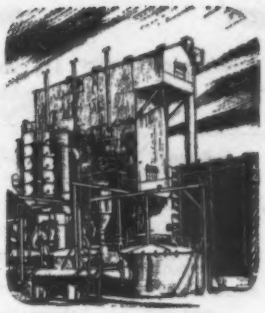

Continuous Casting of High-speed Steel

By B. H. C. Waters, M.A., Ph.D., W. H. Pritchard, A. Braybrook, A.Met., A.I.M., and G. T. Harris, M.A., F.Inst.P., F.I.M.

SYNOPSIS

This paper describes a continuous-casting plant operated on a pilot scale for the production of high-speed steel and other alloy steels as 4-in. square cast billets.

A general description of the plant is given, followed by a description of the steelmaking and metal-handling techniques required for continuous casting. Types of billet fracture encountered in the operation are discussed, together with the pouring and mould systems used to overcome them. An outline of the operational procedure for the plant is then followed by an account of the structure and quality of billets produced on the plant. 1638

INTRODUCTION

IN 1953 a continuous-casting plant was installed at the works of William Jessop and Sons, Ltd. by the engineers Campbell, Gifford and Morton Ltd., based on ideas put forward by B.I.S.R.A.

The project was sponsored by a group of 11 steel-manufacturing companies, to study continuous casting of special steels with particular reference to 18–4–1 high-speed steel. The project was to be carried out over two years and it was expected that steels of superior quality would result using this method or, because of a higher yield, steel would be produced more cheaply than that made by conventional practice.

The plant was designed to produce a 4-in. square, 4-ft long billet, as this was a useful size for handling in subsequent rolling or hammer-cogging operations and could be directly compared with conventional 4-in. ingots. The first cast was made 30th April, 1954 and in 2½ years 200 casts were made. Some of these failed owing to plant imperfections in the earlier casts, but in the last 100 casts only about 5% were unsuccessful due to mechanical faults as distinct from tundish problems and metallurgical quality.

GENERAL DESCRIPTION OF PLANT

The plant was built in an existing shop. This restricted the headroom available and to compensate for this, part of the plant was sunk below floor level in a pit. The height of the plant from the bottom of the pit to the crane track was 37 ft and from ground level to crane track 25 ft. A sketch of the plant is given in Fig. 1. Modifications such as billet roller-guides and electromagnetic stirrers were added later. The total cost of the plant was about £30,000, including modifications.

Two types of pouring method were used to empty a 1½-ton ladle. In one, a lip-poured ladle was used, necessitating a ladle-cradle tilting about the spout as axis: in the other, a bottom-poured ladle was used, suspended from a crane. Leaving the ladle, the metal flowed through a splitter or a tundish to the casting mould or moulds. Originally the plant was designed for two-strand operation but latterly only one strand has been used. From the mould, the cast billet was led through guide rolls down to the main drive rolls and thence to the cut-off station. Here it was cut into lengths which fell into the pit on to a conveyor and were then returned to floor level. Sequential operation of the various parts of the plant was achieved by means of a series of electrical interlock devices.

Parts of the equipment such as pouring and mould equipment whose design was specifically related to a continuous-casting process are described later. Equipment items of a more general engineering character will be described here.

The first of these was the dummy billet design. The joints of the dummy billets were as shown in Fig. 2. Each section was 4 ft long and the original assembled billet consisted of four sections. The lower three of these were hollow, and were removed by hand during casting. The top section was solid and was removed by cutting off the attached section of cast billet. The joints between these billets were not rigid, movement between different billet lengths being restricted to $< 2°$ by adjustment of the stops.

The main withdrawal rolls followed a design basically similar to that of a rolling mill, but with much slower operation. Two sets of two rolls each 8 in. dia., were driven by a 5-h.p. motor through reduction gearing to give a speed variable between about 0·5 and 2·5 rev/min. This gave a casting speed range

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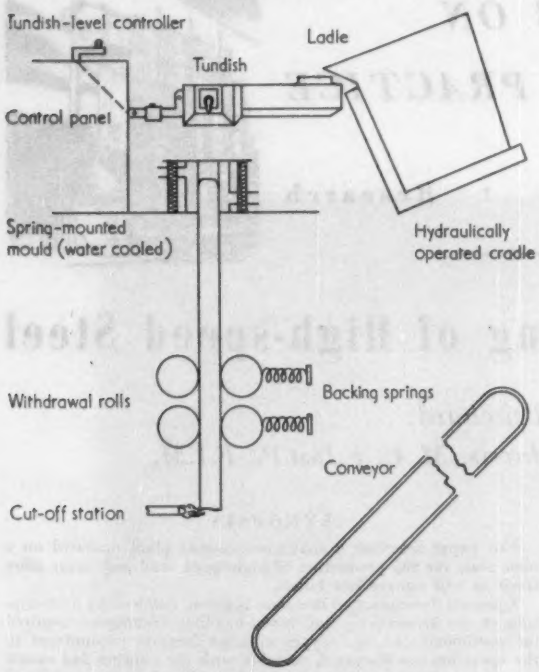


Fig. 1—Continuous-casting plant as erected (not to scale)

of 10–60 in./min. The roll material was mild steel; a heat-resistant steel was tried for one roll and was satisfactory, but not necessary. Back-up pressure was applied on one roll of each set. This was variable but was usually set at 3000 lb. The movable roll rested against a stop so that the space between the rolls was $\frac{1}{8}$ in. less than the dummy-billet section. The rolls were driven through wobblers in such a way that they acted continuously against a restraining brake. Thus erratic withdrawal due to backlash was prevented. On two-strand operation the rolls withdrawing both billets were coupled together. This was undesirable as it was not possible to stop one strand, for instance to heal a fracture, without interfering with the operation of the other strand. Ideally for multi-strand operation, each strand should have its own set of withdrawal rolls, independently driven.

Two sets of guide rolls were also used on the plant, each set with a different purpose. The top set ensured final alignment of the top dummy-bar section as it entered the mould, and also supported and aligned the cast billet immediately below the mould. This set consisted of two groups of four rolls each. The top rolls were spring-loaded with springs of 300-lb/in. rating and they were pre-tensioned to 15 lb whilst the bottom four rolls had springs of 500-lb/in. rating pre-tensioned to 25 lb. The rolls were water-cooled by water running over their faces, and their design is shown in Fig. 3. The bottom set of rolls was used to guide and align the dummy billets as they were fed up into the main drive rolls before a cast. This set consisted of three pairs of rolls one above another.

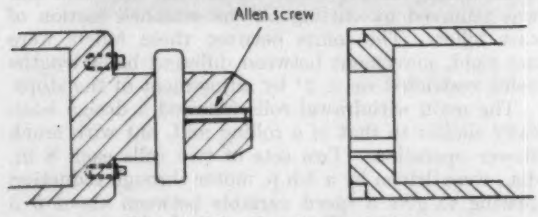


Fig. 2—Joint for dummy billet

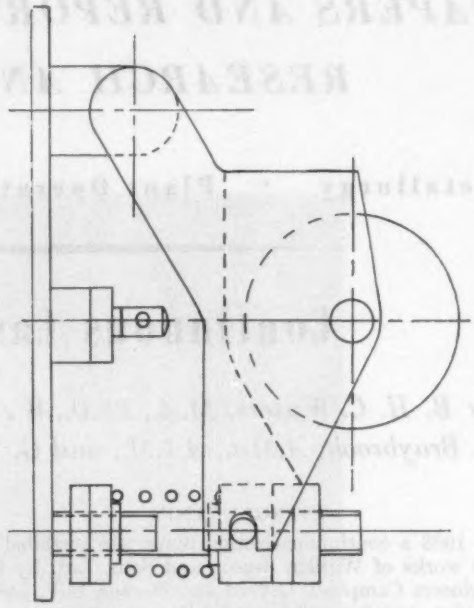


Fig. 3—Design of top set of guide rolls

These rolls were also spring-loaded, with springs of higher rating and with greater pre-tensioning than those of the top set of rolls. This was because they had to guide the billet against any off-centring force of the drive rolls. The spring ratings of the bottom set of rolls were: top two rolls 1000 lb/in., bottom four rolls 2000 lb/in., and they were pre-tensioned to 50 and 100 lb respectively. These rolls were water-cooled by internal water-cooling of the spindle. Their design is shown in Fig. 4.

In some of the work the billet surface was spray-cooled. For this water was usually sprayed at mains pressure of 100 lb/in^2 through atomizing nozzles. These sprays were set immediately below the mould and then down the plant for a distance of about 6 ft.

Cutting off the cast billet was accomplished by means of standard oxy-propane burners using iron-

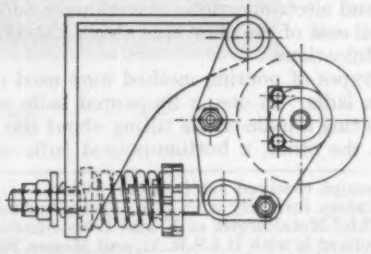


Fig. 4—Design of bottom set of guide rolls

Table I
CUT-OFF TIMES AND TORCH OPERATING DATA FOR VARIOUS STEELS

Steel Quality	Minimum Cut-off Time for 4-in. Billet, s
En 58A 18–8 stainless	40
18–4–1 high-speed	55
Austenitic heat-resisting steel (Jessop R 22)	55
En 56D 13% Cr steel	50
En 59 valve steel	65

Note: Cutting pressure (100 lb), heating pressure (110 lb), propane pressure (2 lb), and powder dispenser (12 lb) were all constant throughout.

powder injection. These burners were set on a carriage which traversed horizontally and moved downwards with the cast billet, being clamped to the billet by means of pneumatically operated clamps. The clamps and burners were operated by means of limit switches and electrical relays. Details of cut-off speeds and powder-flow rates are given in Table I. They depended on the steel quality cast and the billet temperature. By means of these burners the billet was cut into 4-ft lengths.

These lengths then fell into a basket which was counter-balanced so that as the billet entered, the basket tilted forward and lay against a conveyor. The conveyor then picked up the billet, engaging a dog with the bottom of the cast billet, and took it to floor level where it was removed into a pan by means of a hydraulically operated ejection basket. Electrical interlocks again secured the automatic operation of the different components of this disposal system at the correct times.

STEELMAKING FOR CONTINUOUS CASTING

In this work a variety of steels was continuously cast, and typical analyses of different steels cast are given in Table II. Most of the work was, however, done with 18–4–1 high-speed steel, although many casts were made with 18–8 stainless steel. Most of the steelmaking techniques were the same as those used in standard practice. The steel was made in an obsolescent type of 1½ ton bottom-return basic-arc furnace having 8-in. electrodes and a transformer rating of 575 kVA.

One feature of continuous casting was that metal had to be tapped hotter than in standard practice because teeming times and rates were normally longer than standard. Typical tapping temperatures for different casting speeds are given in Table III where tapping temperatures for standard ingot practice are also given. It will be seen that, in general, metal for continuous casting was tapped some 40° C hotter than in standard practice, the temperature being taken by a dip couple in the furnace immediately before tapping and maintained to ± 10° C. Tapping metal too cold caused snottering and a poor lapped surface, particularly with 18–8 steel, and tapping too hot gave a shelly surface and sometimes caused a bleed. It also slightly increased centreline porosity, although in this respect it was secondary to the effect of casting speed.

Table III
TAPPING TEMPERATURES FOR VARIOUS CASTING SPEEDS

Steel Type	Casting Speed, in./min	Tapping Temperature, °C
18–4–1 high-speed	12	1540–1550
	24	1530–1540
	36	1520–1530
	Conventional casting	1460–1480
18–8 stainless	24	1600–1620
	36	1580–1600
	48	1560–1580
	Conventional casting	1500–1520

The gas content of the steel was particularly important in continuous casting. High tapping temperatures made the more simple steels prone to hydrogen porosity so that care had to be taken to ensure a good boil and dry alloy additions. A more complex requirement, however, resulted from the effect on the casting stream of alumina. If aluminium deoxidation was used, alumina deposition was caused on the small meter nozzles that controlled the flow rates and this interfered with metal flow. Therefore calcium–silico-manganese was used as a deoxidant, and this prevented alumina deposition in most of the steel qualities cast. However, alumina deposition still occurred with 18–4–1 steel even if aluminium deoxidation

Table II
TYPICAL ANALYSES OF STEELS CAST

Steel Type	En No.	Jessop Code No.	Analysis, %						
			C	Si	Ni	Cr	W	V	Mo
Plain carbon	9	B1	0·55	...	...	...	...	...	...
Valve steel	59	H29	0·80	2·00	1·50	19·5	...	...	...
18–4–1 High-speed	...	J3	0·76	...	...	4·25	18·0	1·40	...
18–8 Stainless	58A	R3	0·12	0·30	10·50	17·0	...	...	...
Austenitic heat-resisting	...	R22	0·21	1·00	11·5	22·5	2·75	...	...
13% Chrome	56D	R34	0·30	...	...	13·0	...	...	...
1¼% Ni–Cr–Mo	24	G5 spl	0·32	...	1·55	1·10	...	...	0·20

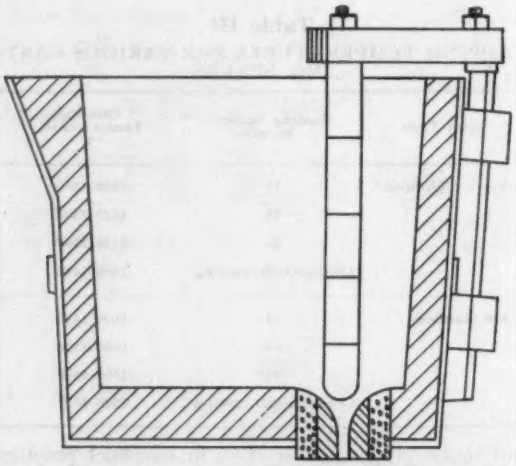


Fig. 5—Bottom-poured ladle used for continuous casting

additions were avoided, probably due to the occurrence of aluminium in the scrap and alloy additions used. Efforts were made to flush out the alumina by argon injection but these were not successful. In the end the pouring system had to be designed and casting speeds adjusted so as to work satisfactorily despite the occurrence of alumina deposition at nozzles.

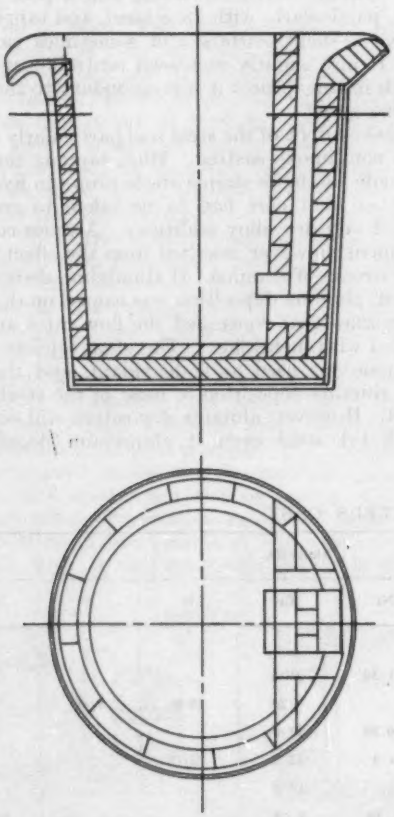


Fig. 6—Lip-poured ladle used for continuous casting

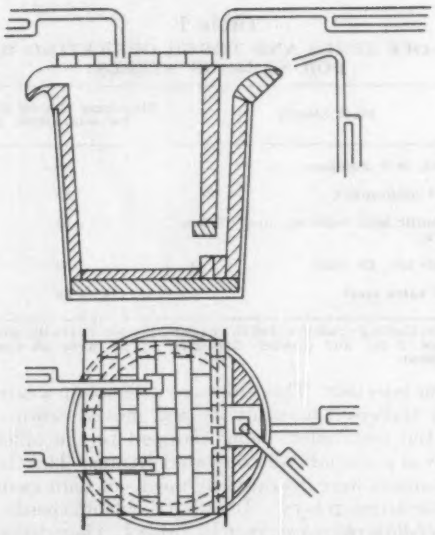


Fig. 7—Arrangement of burners for heating lip-poured ladle

Two types of ladle were used in the course of this work. One was bottom-poured and the other was lip-poured. The bottom-poured ladle was preheated to about 900° C in accordance with standard practice, but the lip-poured ladle was heated to 1350–1400° C so as to minimize the chill of the ladle on the liquid steel.

LADLE-POURING SYSTEM

More severe demands are made on a continuous-casting ladle than on standard ladles because metal must be delivered at a slower rate than normal over a longer time than normal, casting times of 30–60 min being customary.

The standard ingot casting ladle used for bottom pouring in continuous casting is shown in Fig. 5. The refractories used to line the walls were of medium alumina firebrick, 3 in. thick, backed with about 1 in. of ganister. The bottom consisted of 4½-in. high alumina firebrick with a medium alumina nozzle rammed in place with refractory composition (' compo '). The stopper and sleeves were also of medium alumina firebrick. The preheat temperature for this ladle was 900° C. A sketch of the lip-poured

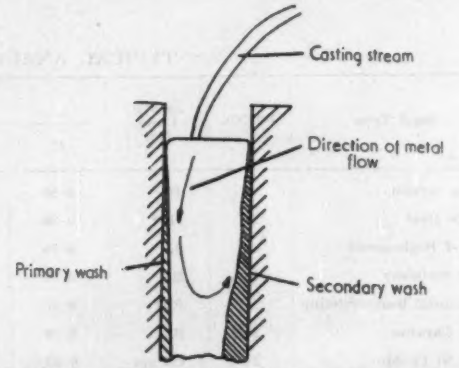


Fig. 8—Metal wash of billet skin during casting

ladle used in this work is given in Fig. 6, and this ladle was specially designed for continuous casting. The walls were of high-alumina firebrick and both walls and bottom were of 4½-in. firebrick backed with about 1 in. of ganister. The preheat temperature of this ladle was 1350–1400° C, and this is why higher alumina firebricks were used and the lining thickness increased to conserve heat.

Preheating of both types of ladle was by burners using an air–gas mixture using air at 2–3 in. water gauge, and town gas (calorific value 500 B.t.u./ft³) at 4–5 in. water gauge. All ladles were heated in the upright position. The bottom-poured ladles were heated by a single burner in the ladle body, together with a smaller burner at the nozzle. The top of the ladle was covered with mild-steel sheet, about 16 gauge. The lip-poured ladle required two body burners and two burners for the lip and spout as shown in Fig. 7.

Fig. 9—Billet failure due to metal wash

Fig. 10—Sectioned billet skin showing metal wash × ½

The lip-poured ladle was better for casting metals which tended to erode firebrick refractories, and it had the advantage of avoiding any danger of a running stopper. Also, for larger installations, lip-poured ladles could move easily and be continuously heated throughout casting. Bottom-poured ladles gave reduced refractory and fuel costs since they were not preheated so much as the lip-poured ladles. Plant construction would be simplified if bottom-poured ladles were used because no ladle-tilting mechanism would be needed; bottom-poured ladles gave improved slag separation and, also, if a cast had to be stopped the metal from a bottom-poured ladle could be cast into standard ingot moulds. No great trouble was experienced from running stoppers with bottom-poured ladles: in over 150 casts with such a ladle only one running stopper occurred. Metal from this cast was dumped in a slag pan placed about 3 ft from the pouring station to deal with such an emergency. Thus, it is considered that for smaller quantities of metal such as were cast in this work, a bottom-poured ladle is better: lip-poured ladles might have advantages for larger quantities of metal when freedom from running stoppers and increased ease of continuous

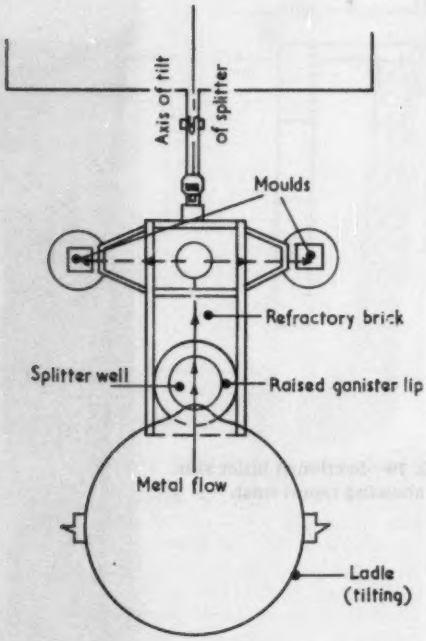


Fig. 11—Splitter design and operation

ladle-heating throughout casting would outweigh the advantages of bottom pouring.

TYPES OF BILLET FRACTURE

The most troublesome failure experienced in continuous casting was that which involved breakout of liquid metal through the solid envelope immediately below the casting mould. This could be due either to rupture of the envelope within the mould, and failure of the rupture to heal, or to remelting of the solid envelope.

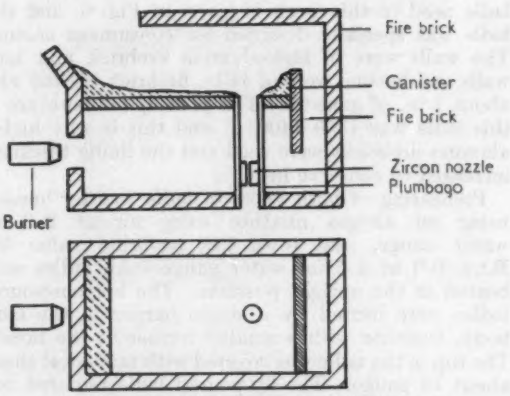


Fig. 13—Original tundish design

The mechanism by which rupture occurs has been extensively described.[1] Briefly, it is caused by mould friction which restrains the top of the solid envelope while the bottom is pulled down. A hot tear results and if it is not healed the liquid metal can escape from the bottom of the mould when the torn section is withdrawn. The mould and withdrawal system, described later, were designed to avoid this kind of failure and it only occurred when they did not work properly.

The second type of rupture resulted from remelting owing to the metal being cast too hot or, much more often, to non-axial pouring resulting in wash on the solid metal skin. The first type of remelting was avoided by careful specification of the tapping temperature, as already described. Remelting by washing

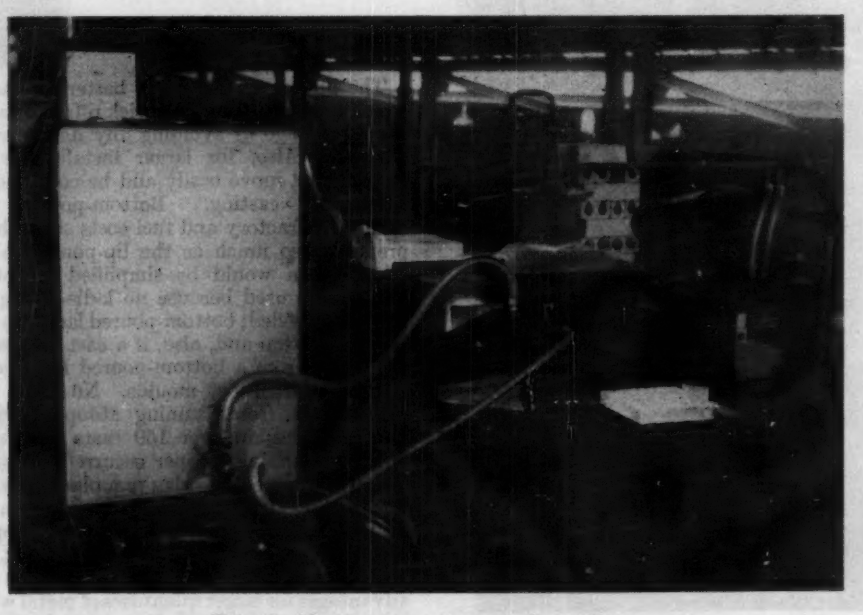


Fig. 12—Heating arrangements for splitter

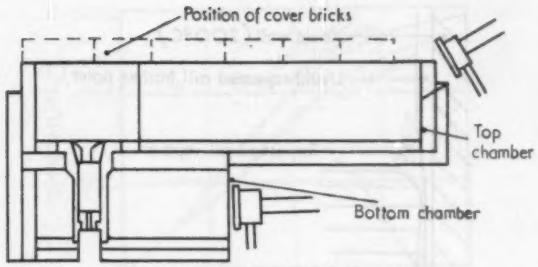


Fig. 14—Final tundish design

took place either by direct impingement (primary wash, Fig. 8) or by wash from a deflected stream (secondary wash, Fig. 8). A failure caused by wash is shown in Fig. 9 and a photograph of a sectioned billet shell in which wash has occurred is shown in Fig. 10. If remelting occurred in the mould the metal did not run out entirely, but escaped between the solid envelope and the mould wall and solidified there. If this happened, a shelly billet surface resulted causing considerable mould wear.

From this it will be seen that to prevent billet fracture a suitable design of mould and withdrawal system is required, together with a pouring system which gives axial entry of liquid metal into the casting mould so as to avoid wash. In addition, correct tapping and casting temperatures are needed.

POURING SYSTEM

An integral part of a continuous-casting plant is equipment for conducting liquid metal from the ladle to the continuous-casting mould. In this work two conventional methods were used: (*a*) splitter pouring, and (*b*) tundish pouring.

Splitter pouring was used first and, although it was subsequently discarded because the non-axial pouring associated with it gave metal wash, its use is described here since it has advantages and in casting elongated sections, such as slabs or ellipses, pouring along the major axis is likely to avoid problems of wash.

Splitter pouring was used in two-strand casting, and a sketch of the splitter showing its mode of operation is given in Fig. 11. Medium-alumina firebricks were used as the refractory, and the splitter was preheated internally to 1400° C (thermocouple) and at the lips to 1500° C (optical pyrometer) before use. To do this, three Selas burners were used, situated as shown in Fig. 12. These burners operated on town gas at 4–5 in. water gauge and compressed air at 100 lb/in². The splitter design was such that the well at the back was kept as small as possible to minimize radiation losses, while for the same reason a straw cover was kept on the metal there during casting. In addition a slag weir was introduced between the splitter well and side arms, to minimize slag entrainment.

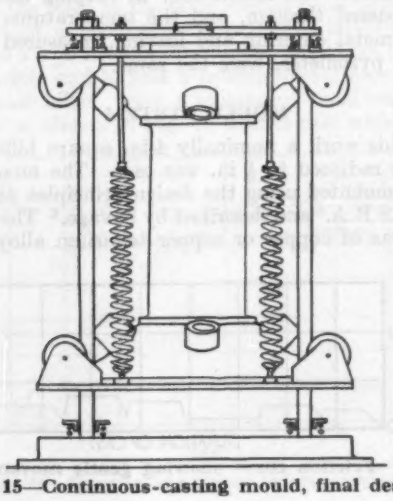


Fig. 15—Continuous-casting mould, final design

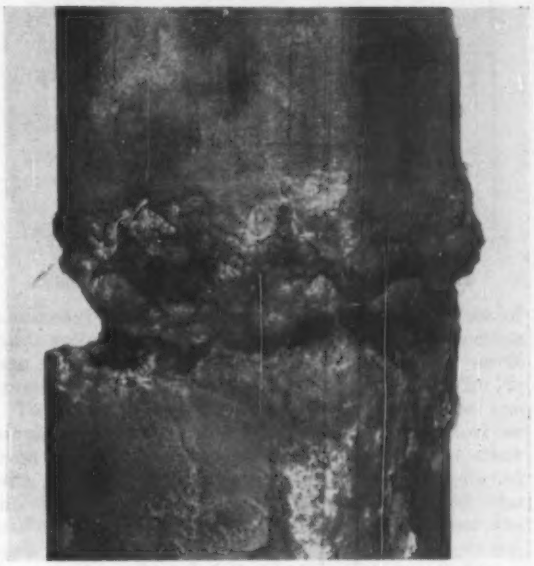


Fig. 16—Billet failure due to bad alignment

Tundish pouring replaced splitter pouring primarily because it centralized the metal stream to the mould. Tundish pouring was only used with single-strand operation because casting with two coupled strands gave casting problems mainly due to the coupling of the two roll systems as previously described. Considerable development work was done on the tundish, for which the general requirements were as follows:

(1) A low throughput rate was required, so that a casting rate as low as 2 tons/h could be achieved
(2) It was necessary for the metal stream to be visible to the operator at a point about 8 in. below the tundish so that he could see the surface of the metal in the mould and control the level
(3) Reliability of the tundish was necessary
(4) It was desirable to provide a means of regulating the metal flow from the tundish.

The tundish first used was based on a B.I.S.R.A. tundish[2] shown in Fig. 13. Certain weaknesses revealed under works conditions were the fragile nature of the downtube, inadequate metal capacity, skulling of the metal surface in the tundish, making up of the small meter nozzle, and leaking of metal through the tundish refractories. As a result of development work the tundish shown in Fig. 14 was evolved. This tundish was basically made of high-alumina firebrick, and the metal entered a bricked

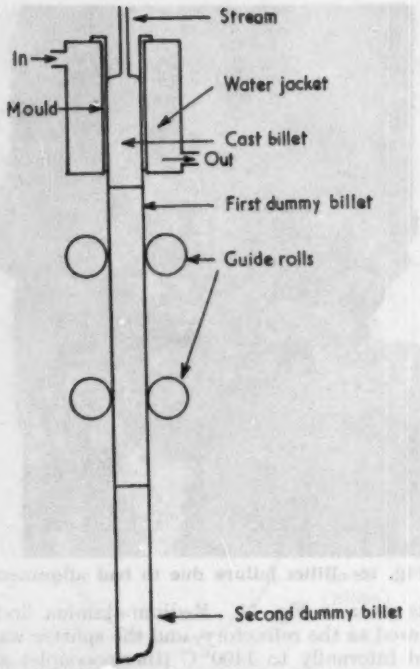


Fig. 17—Location of top guide rolls

body, at one end of which a downtube was set. The body was such that it could contain a depth of 5 in. of metal, so that the total metal content of the tundish was about 2 cwt. The downtube was rammed in place using a refractory composition rammed against a sillimanite hearth-brick which separated the top liquid-metal chamber from the lower heating chamber. At one end of the tundish, a flue led gases from the bottom heating chamber to atmosphere. The downtube which passed down through the heating chamber and which carried the small meter nozzle set in sillimanite cement to control the metal flow, was made of plumbago and was greatly thickened to 1¼-in. thick walls as compared with the original downtube walls (¼ in. thick). Choice of the meter nozzle material and design of the meter nozzle were the subject of much work to overcome the blocking effect of the alumina which was deposited on it when 18–4–1 high-speed steel was cast. Experiments were conducted in which the nozzle was made in a material to which it was hoped alumina would not stick either because of non-reaction or because highly fluid reaction products would form and be washed away. Work was also done on modification of the nozzle contour to try to minimize turbulence and, by promoting a smooth flow, to minimize alumina deposition. This work was not successful and in the end a zircon meter nozzle with a hole drilled straight through it similar to that used on other continuous-casting plants was adopted as standard. It was, however, made larger than was necessary for the metal flow required so that alumina deposition did not reduce the metal flow rate below that required. Despite this, with 18–4–1 steel it was not possible to cast satisfactorily at speeds below about 28 in./min and this speed

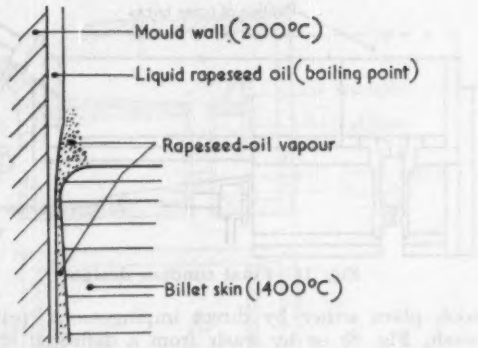


Fig. 18—Mode of lubrication by rapeseed oil

limitation had a great adverse effect on the billet quality produced with this material, as described later. Use of such a nozzle meant that means had to be employed for reducing the metal flow through the nozzle below that which it could accommodate, and to do this a stopper was introduced at the top of the downtube. This seated on a large nozzle, 3 in. outside diameter, 1⅛ in. bore, set in the top of the downtube with sillimanite cement.

The stopper end was made of plumbago to minimize oxidation attack on the stopper rod during preheating, while the stopper rod sleeves were of medium-alumina firebrick. The whole of the tundish and stopper rod assembly had to be bricked very carefully, as any inattention in bricking resulted in leaking during casting.

This tundish was preheated by means of two Selas burners and was continuously heated during casting by the same burners. One burner was set to heat the liquid metal chamber and impinge at 30° to the surface of the metal there throughout casting. The other burner heated the downtube chamber, and was also kept on throughout casting. Before casting the tundish was heated to 1300–1350° C in the top chamber and 1450–1500° C in the downtube, temperatures being measured by rare-metal thermocouples. This tundish was satisfactory in keeping metal hot as it passed through, and the temperatures of the liquid metal entering and leaving, measured by an optical pyrometer, were the same.

MOULD SYSTEM

In this work a nominally 4-in. square billet with corners radiused to ½ in. was cast. The mould was spring-mounted using the design principles patented by B.I.S.R.A.[3] and described by Savage.[4] The mould tube was of copper or copper–tellurium alloy, ¼ in.

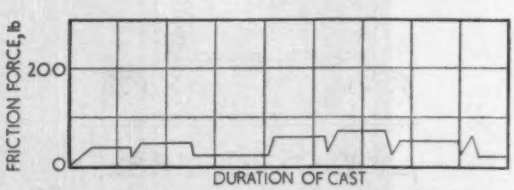


Fig. 19—Friction curve showing gentle movement of mould

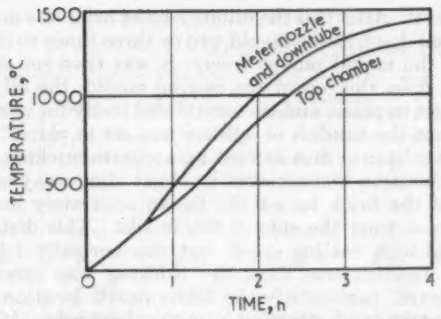


Fig. 20—Heating of tundish before casting

thick, with an annular water jacket of ½ in. through which water passed at 150–250 gal/min, being heated by 2–3° C during passage. The mould was 19 in. long giving a working length of about 15 in., since the meniscus of liquid metal in the mould was normally about 4 in. from the top during casting.

The first mould system used compression springs with a rating of 200 lb/in. suspended on a metal–rubber bush system. This was replaced by a mould shown in Fig. 15. A feature of this mould was that it used extension rather than compression springs of the same rating as the other mould, and by their action they automatically tended to centre the mould during casting. Lateral movement of the mould was restrained to ⅛ in. in any direction by means of four guide posts. A further feature was the provision of compression springs which acted to cushion the movement of the mould at the extreme ends of its travel either downwards or upwards. Thus, when the mould approached full deflection, an increased restraining force came into play and if the mould was suddenly released from extreme deflection the top set of springs cushioned its movement on return.

Of equal importance to mould design was, however, the provision and correct location of spring-loaded guide rolls immediately below the mould. These have already been described: their function was to locate correctly the dummy billet as it entered and left the casting mould and also to support the dummy billet. Thus, if the force in the billet moved from tension to compression, as could occur immediately following the release of a high friction, the dummy would not buckle at the joints and impose a sudden lateral movement on the cast billet. Such a movement could result in a failure as shown in Fig. 16 from a cast which was

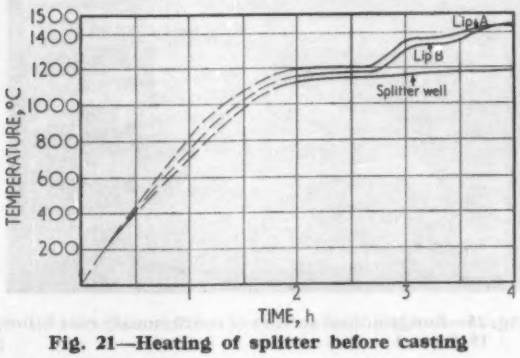


Fig. 21—Heating of splitter before casting

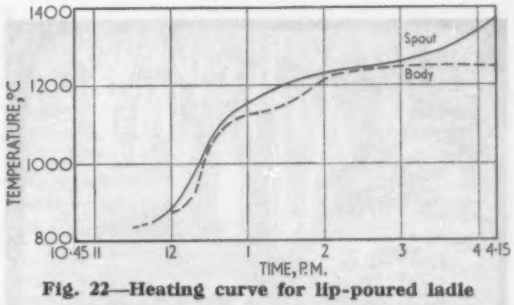


Fig. 22—Heating curve for lip-poured ladle

conducted without guide rolls. Thus the top set of guide rolls was so placed that the top dummy section was completely located by guide rolls both as it entered and left the casting mould, as shown in Fig. 17.

To assist casting it was desirable to reduce the mould friction as much as possible. Such friction arose from slight mould bowing and undulations together with small pits formed by splash particles being forced into the mould wall. For this reason the mould tube was made nominally parallel but any slight taper was such that the bottom end of the tube was the wider. It was cleaned and dressed, if necessary, before a cast, with a block file, the filing again being arranged so as to widen the bottom end. The effect of minor surface imperfections was overcome by suitable lubrication using a volatile lubricant such as rapeseed oil or water, so as to maintain a vapour film between the cast billet surface and the mould surface, as shown in Fig. 18. An indication of the efficiency of the dressing and lubrication of the mould tube was provided by the visible vertical movement of the mould assembly during casting. Ideally it was small and occurred very gently giving a so-called friction curve as shown in Fig. 19.

The lubricant used was also important in providing a reducing environment above the metal in the mould. Many investigators[5, 6, 7] consider the provision for such an environment important, to reduce oxide formation, although this was not substantiated in the present work.

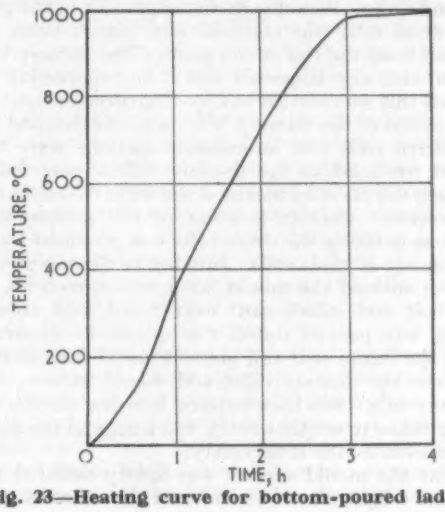


Fig. 23—Heating curve for bottom-poured ladle

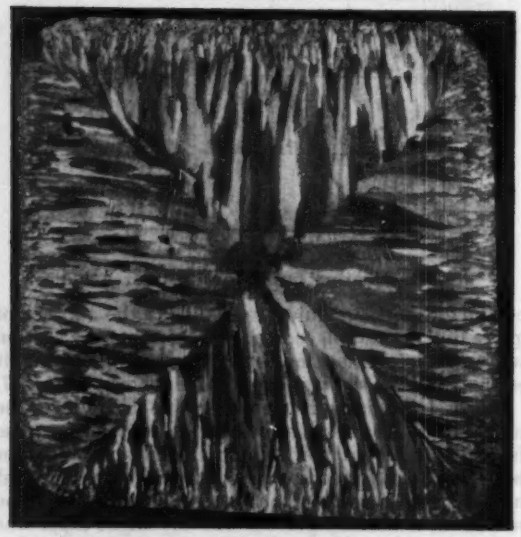


Fig. 24—Transverse section of continuously cast billet, 18–8 steel × ¾

Summarizing, the mould system used in this work consisted of a spring-mounted mould used in conjunction with support and guiding of the dummy and cast billets below the mould. The mould was adequately lubricated and its contour was so designed that, although nominally parallel, any slight divergence from this condition was such as to provide a wider outlet at the bottom than the top.

PLANT OPERATION

Brief details of the operation of the continuous-casting plant are given here, bringing together the items mentioned previously.

In setting up the plant for a cast the mould surface was first cleaned and inspected visually. Local dressing was then carried out if necessary, as it occasionally was, and the mould dimensions were then checked with a micrometer to ensure that any taper was big-end down. Periodically the alignment of the guide and drive rolls was checked with plumb bobs, suspended from the top of the plant. The dummy-billet joints were also inspected and if any tightening was needed this was carried out by stop adjustment. The top section of the dummy billet was then hoisted into the drive rolls and subsequent sections were then added by hand as the dummy billets were fed up through the plant by means of the drive rolls operating in reverse. Correct location of the dummy-billet sections entering the drive rolls was provided by the bottom set of guide rolls. Just before the top dummy section entered the mould, as it was moved up, the key bolt and splash can[8] were fitted, and asbestos string was packed round the splash-can groove to form the initial seal and prevent metal flash forming between the dummy billet and mould surface. The dummy billet was then entered into the mould, care being taken to see that entry was true, and the mould was moved a little if necessary.

Next the mould surface was lightly smeared with oil, and a light film of oil run over the asbestos string to wet it. After this the dummy-billet head was moved up and down in the mould two or three times to check that the mould moved freely: it was then set about 6 in. from the top of the casting mould, the oil ring was set in place, and the mould was ready for casting.

Next the tundish or splitter was set in place. The splitter lip was first located by accurate bricking, but final location was achieved by slight filing away of the lip of the brick to set the lip an accurately known distance from the side of the mould. This distance varied with casting speed, but was normally $1\frac{1}{16}$ in. If a tundish was used the bricking was carefully inspected, particularly the meter-nozzle location and the setting and ramming-in of the downtube. If this was satisfactory the tundish was set over the mould, being located with a fine rod passed vertically down through the meter nozzle. Heating of the tundish or splitter was then begun. Temperatures were taken periodically during heating and at first the heating was slow. Typical heating cycles for the tundish and the splitter are shown in Figs. 20 and 21. It was necessary that the final temperature required should, in each case, be maintained for 60 min before casting.

At the same time the ladle heating was begun and again control was kept on the heating, typical heating curves for lip- and bottom-poured ladles being shown in Figs. 22 and 23. Again it was necessary that the final temperature should be maintained for 60 min before casting. While this was going on the cut-off was checked, the powder in the dispenser being emptied and refilled before each cast to prevent clogging.

Finally the metal was tapped at a specified temperature, the ladle of metal taken to the plant, and casting began. At the start of casting an oil can was normally used for auxiliary lubrication. Routine maintenance was carried out to schedule.

The plant was critical in operation in that failure

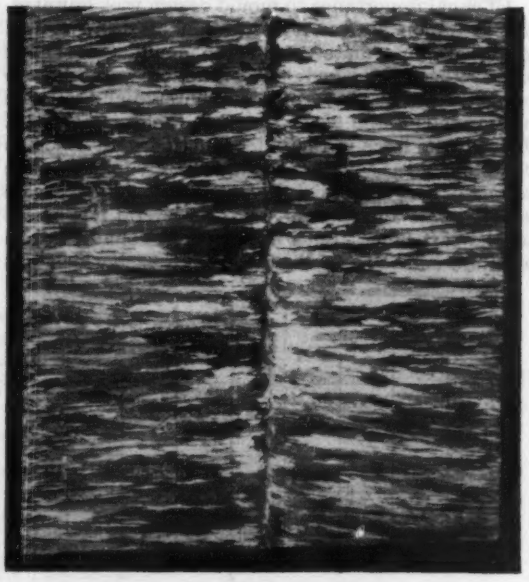


Fig. 25—Longitudinal section of continuously cast billet, 18–8 steel × ¾

Table IV

SEGREGATION IN CONTINUOUSLY CAST STEELS

Element	En58A 18–8 Stainless		Austenitic Heat-resisting (R22)		En59 Valve Steel		En56D 13%Cr		18–4–1 High-speed	
	Mean	Range	Mean	Range	Mean	Range	Mean	Range	Mean	Range
C	0·27	0·01	0·25	0·02	0·78	0·06	0·30	0·03	0·74	0·02
Mn	0·61	0	0·78	0·03	0·48	0·02	0·45	0·03	0·36	0·01
Si	0·61	0·04	1·30	0·12	2·21	0·06	0·30	0·01	0·36	0·02
S	0·013	0·004	0·011	0·003	0·010	0·001	0·019	0·006	0·010	0
P	0·013	0·001	0·020	0·002	0·019	0·006	0·024	0·003	0·019	0·004
Ni	9·29	0·12	11·7	0·3	1·48	0·04	0·32	0·03	...	...
Cr	15·50	0·12	22·9	0	19·6	0·3	12·9	0·4	4·40	0·08
W	...	...	2·36	0·44	...	...	...	...	18·70	0·22
V	...	...	...	...	...	...	...	...	1·44	0·05
Mo	...	...	...	...	...	...	...	...	0·24	0·01

of a single item of equipment could often stop a cast: particular attention was paid to the electrical interlock system and limit switches used.

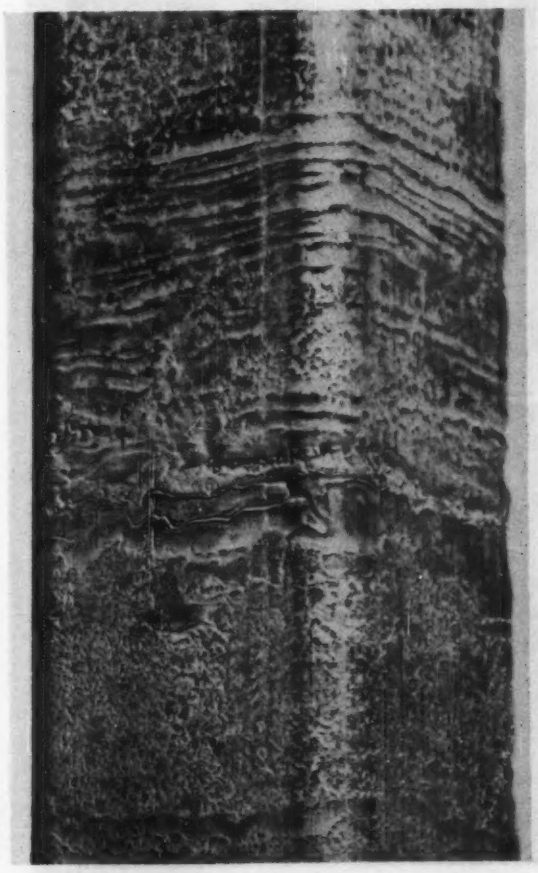


Fig. 26—Surface of billet showing scars due to hot tears healed in the mould

BILLET QUALITY

With 18–4–1 high-speed steel it was not possible to get a satisfactory structure of the billet produced at casting speeds (e.g. 36 in./min) at which the plant functioned satisfactorily with the billet section used. Only at slow casting speeds of about 12 in./min was satisfactory billet quality obtained and at these speeds the metal-handling problems were such that the tundish nozzle made up after about 10 cwt metal was cast (about 20 min).

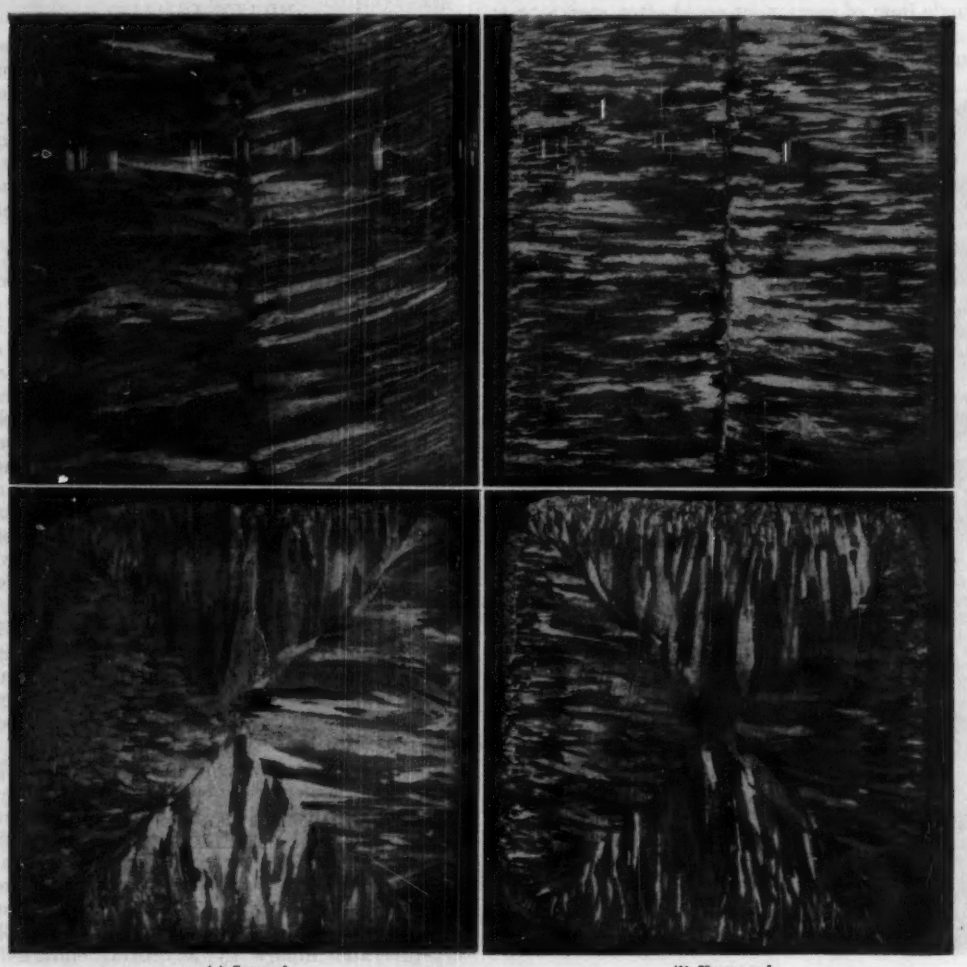
Although casting 18–4–1 steel was the first object of the plant, several other steel qualities were cast. For all qualities the structure of the product as cast was like an ingot rather than a worked billet. Macrographs of transverse and longitudinal sections of such continuously cast billet are shown in Figs. 24 and 25. Details of the compositions of qualities cast are given in Table II. All the billets were typically rhombic in section, being off-square by about 0·1 in. often with some sides slightly concave. All the billets had some degree of axial porosity as discussed later, the character and severity varying with the type of metal cast. Sometimes, as in 18–8 steel (R3), it took the form of a string of voids situated at the billet axis, while in other alloys such as valve steel (H29) the porosity consisted of a general looseness similar to 18–4–1 steel (J3). Severity of the porosity varied with the alloys: thus one quality of heat-resistant austenitic steel (R22) showed much less porosity than other steels cast. As a general rule the porosity found in continuously cast material at the casting speeds used was found to be a little worse than that of a conventionally cast 4-in. ingot, having a taper of $\frac{1}{8}$ in./ft. The billet structure depended on the alloy cast: stainless qualities showed an outer zone of chill crystals followed by columnar crystals which grew transversely and appeared to penetrate to the centre of the section. Other qualities investigated (E9, En59, 18–4–1, En56D, and En24) showed an outer chill crystal zone followed by a zone of columnar crystals and finally by a central equiaxed zone. Billets of En59, En58A, and R22 were examined for segregation by taking drillings for chemical analysis

from about 10 points across the 4-in. square section. Results, given in Table IV, showed that no marked segregation had taken place. Carbide segregation did, however, occur in 18–4–1 steel as described later.

The billet surface obtained in continuous casting varied with the type of metal cast and to a less extent with casting speed, faster speeds giving better surfaces. The formation of the billet surface is such that the contour is affected by the rate of strengthening of the skin immediately after solidification. It therefore depends both on the rate of thickening of the skin and on the strength of the skin near the melting point. If the skin rapidly thickens relative to the downward movement of the billet in the mould the deep corrugations persist: they also persist if the skin strengthens rapidly. Of the two metal qualities continuously cast in quantity during this work, the billet surface of 18–8 stainless steel for similar casting conditions, was worse than that of 18–4–1 high-speed steel. Tundish pouring was also better than splitter pouring because it reduced splashing which gives rise to surface lapping and also because it minimized bleed-outs in the mould which, as described previously, resulted in a shelly surface. Correct mould functioning was necessary for a good surface: the poor surface produced at a hot tear which has subsequently been healed is shown in Fig. 26.

Although grain-size refinement is a major advantage in the continuous casting of non-ferrous metals, such refinement was not found in continuously cast steel as compared with steel statically cast in conventional ingot moulds of similar section. Water-spraying below the mould was tried, but it did not markedly alter the billet structure and it only slightly refined the grain size, or in 18–4–1 steel, the carbide distribution.

A comparison between sprayed and unsprayed material is shown in Fig. 27 for 18–8 stainless steel and in Fig. 28, for 18–4–1 high-speed steel. Electromagnetic stirring below the mould was also tried, to see if it would influence billet structure. Four sets of three-phase 50 cycle a.c. three-pole laminated-yoke stirrers were used, carrying a current of 9 A at 400 V.



(*a*) Sprayed (*b*) Unsprayed

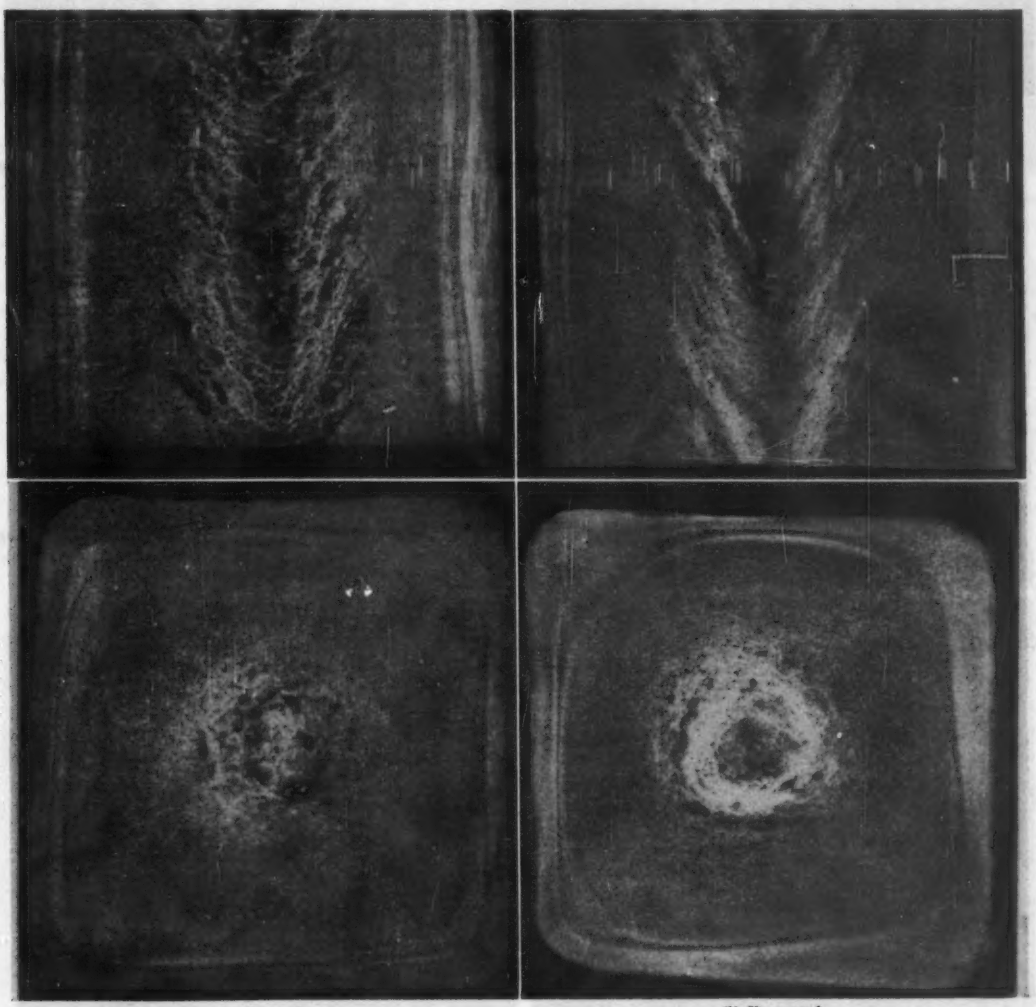
Fig. 27—Comparison of water-sprayed and unsprayed 18–8 billet × ⅔

Each pole had 500 turns of asbestos-covered wire, and a water-cooled austenitic stainless-steel water jacket ¼ in. thick was placed between the billet and poles of the stirrer. These stirrers were placed below the mould at a position embracing the final point of solidification of the billet. They did not refine the structure of 18–4–1 steel, but had erratic and often marked effects on the structure of other steels, the effects being very sensitive to the composition of the steels cast. This is now being investigated.

The character of shrinkage porosity in continuously cast billets depended on the steel quality cast and arose from the way in which the steel froze. Thus steels such as 18–8 which solidified with a highly progressive grain growth formed a central pipe consisting of axial pencil-shaped voids about ⅛ in. across. By contrast, if the steel did not solidify in a highly progressive fashion, the pipe formed tended to be dispersed. Pipe formation was due to bridging across of outcropping dendrites from the sides of the liquid pit isolating pools of liquid which, on freezing, formed shrinkage voids and tended to deposit segregated carbides.

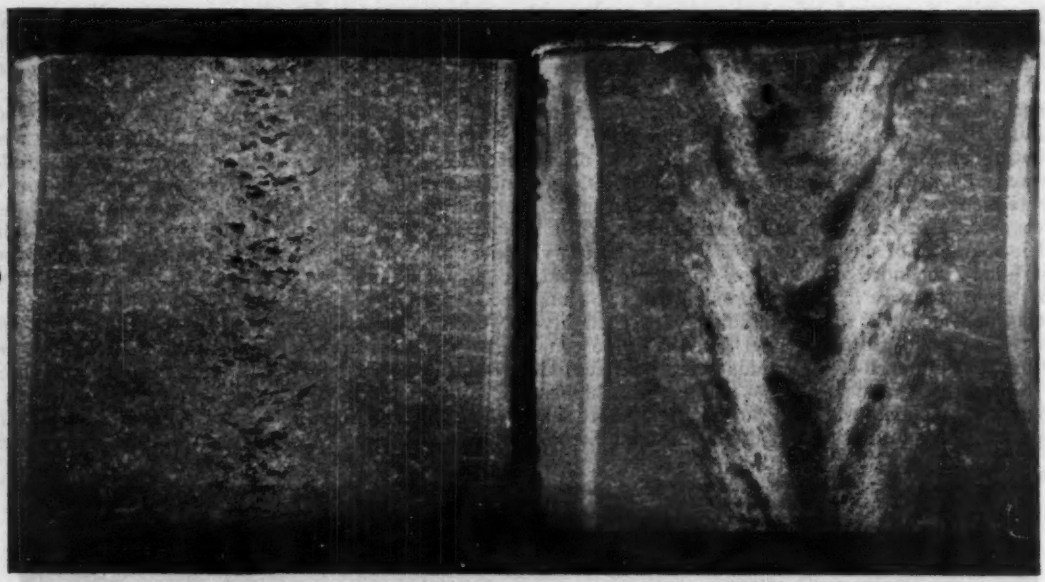
Thus, with shallower liquid pits as were obtained at slower casting speeds, smaller liquid volumes were isolated so that slower casting speeds gave a sounder billet with (in 18–4–1 steel) less carbide segregation, as shown in Fig. 29. It might be expected that water-spray cooling below the mould would have a similar effect in reducing casting-pit depth, but only very slight structural improvement was obtained from water-spraying.

With continuously cast 18–4–1 high-speed steel severe central carbide segregation was found when casting speeds of about 36 in./min, the optimum plant-operating speed, were used. It corresponded to a worsening of shrinkage porosity, both effects being due to the same cause. When casting speeds were



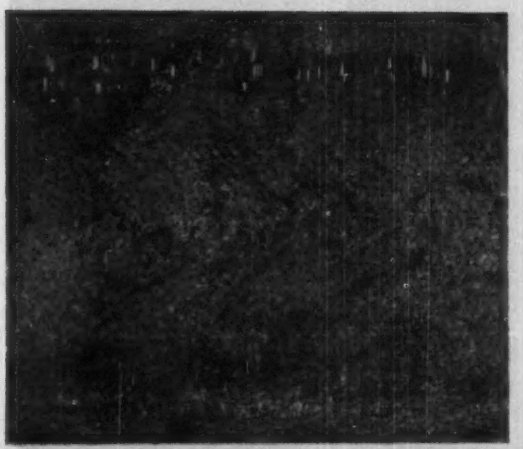
(*a*) Sprayed (*b*) Unsprayed

Fig. 28—Comparison of water-sprayed and unsprayed 18–4–1 billet × ⅔



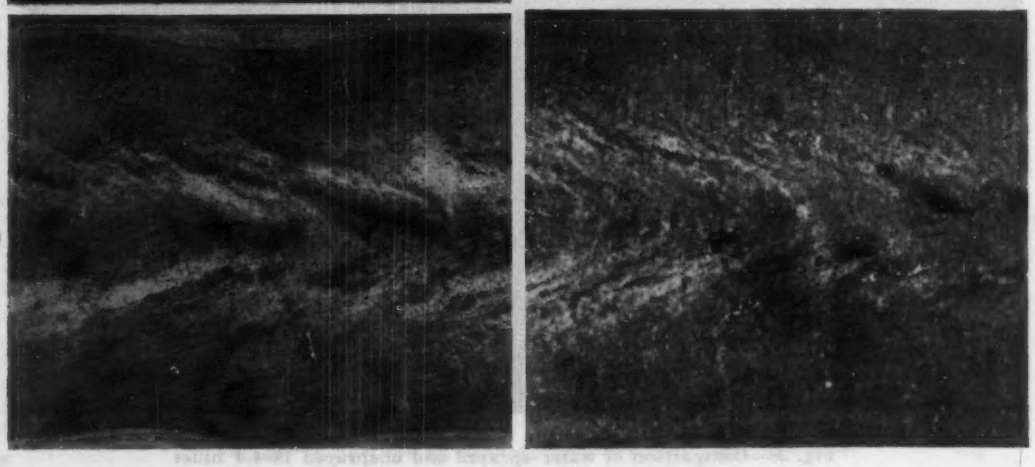
(*a*) 12 in./min (*b*) 36 in./min

Fig. 29—Comparison of billet soundness at different casting speeds, 18–4–1 steel × ⅔



(*a*) 12 in./min (*b*) 24 in./min (*c*) 34 in./min

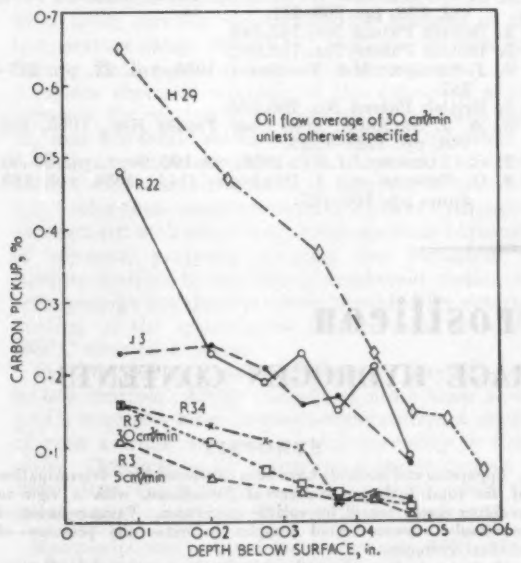
Fig. 30—Carbide segregation in 18–4–1 steel with various casting speeds × ⅔



reduced to 12 in./min the carbide distribution became acceptably fine, but then plant operation problems supervened, as described previously. Comparative carbide structures produced at various casting speeds are shown in Fig. 30.

It was not found that the cleanness of the internal structure of continuously cast steel was inferior to that of standard steel ingots. If anything it was slightly better, although an exact comparison was difficult because the slight differences in cleanness which were looked for could easily have resulted from slight variations in steelmaking from heat to heat. Inclusions found were of silicates or alumino-silicates indicating that they could be derived from the steel itself, or from refractories. In general, billets cast by splitter pouring, with less efficient slag separation, were dirtier than those cast by tundish pouring. Such splitter-poured billets were also more prone to the formation of silicate-type inclusions from the slag, than were the tundish-poured billets. These more frequently contained alumino-silicates with a high proportion of alumina, indicating that they may have resulted from the alumina build-up developed in the meter nozzle being periodically washed through into the continuous-casting mould.

With some simple steel qualities such as En24, special precautions had to be taken to get hydrogen-free material at the furnace, particularly since very hot steel was needed. If these precautions were not taken, the extra hydrogen pickup resulting from the oil used as lubricant in the process gave a wild billet. Such hydrogen pickup was assessed, and for 18–8 stainless steel it was 1·6 cm³ hydrogen per 100 g of steel, when rapeseed oil was used as lubricant and 1·2 cm³ hydrogen per 100 g of steel when water was used as the lubricant. In this test, metal was sampled at the furnace and at the bottom of the plant



Mid-radial Analysis

Quality	J 3	R 34	R 22	R 3	H 29
Carbon, %	0·74	0·31	0·23	0·12	0·80

Fig. 31—Surface carbon pickup in continuously cast steels

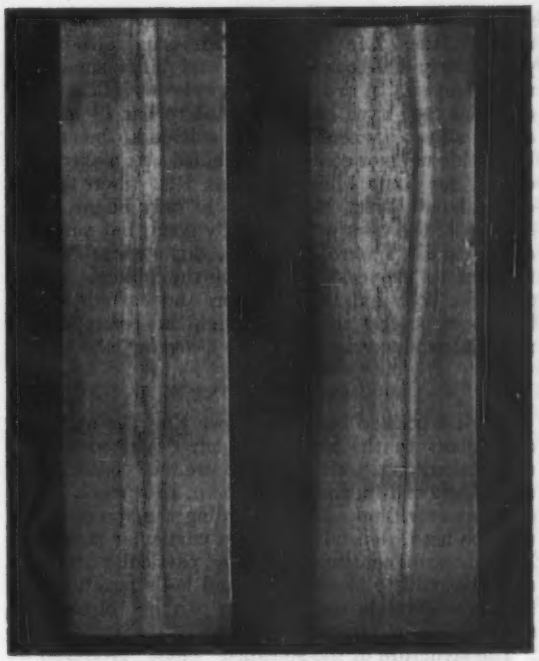


Fig. 32—Banded segregation in bar made from continuously cast 18–4–1 steel, full size

(at about 600° C). Thus, with the alloy steel qualities cast on this plant hydrogen pickup from the lubricant was not a problem: the less highly alloyed steel qualities needed special precautions to ensure a low hydrogen content at the furnace.

Surface carbon pickup, from the lubricant, also occurred in varying amounts with different steels as shown in Fig. 31. The carbon pickup decreased towards the centre of the billet until at about 0·01 in. below the surface the steel had the correct analysis. Such carbon pickup was eliminated by scaling in billet reheating operations, and no trouble from it was encountered with rolled bar.

Billets of all qualities produced in the continuous-casting plant were rolled to 1¼-in. dia. bar (92% reduction) and examined in this condition. In all cases where there was porosity in the original billet this was welded up in the rolled 1¼-in. bar except when the pipe penetrated to the surface at one end. When this happened the surface of the pipe became oxidized and did not weld up: thus the elongated voids in the cast billet gave more trouble from this point of view than small dispersed voids. In no case, however, was the pipe welded up at 66% reduction. At a 92% reduction the cast structure was broken down, but when carbide segregates were present in the original billet these persisted in the rolled bar in the form of heavy banding (Fig. 32). Such banding was very troublesome in billets of 18–4–1 steel cast at fast casting speeds and much of this material was scrapped on account of it. The surface of the rolled 1¼-in. dia. bar has in all cases been comparable to that of standard production, but both continuously cast billets and standard small ingots were normally

surface dressed before rolling down. Normally both required the same amount of dressing, although a particularly bad continuously cast billet sometimes needed excessive dressing. Samples of billets of all qualities cast have been extruded to bar 1¼ in. dia., and in all cases where looseness existed in the original billet this was not completely welded up on extrusion, except for some billets such as R22 where in the original cast billet looseness was only slight. The surface of the bars was generally good, but since the extrusion slugs were machined, this surface bore no relation to the cast surface of the billets. In all qualities both axial segregation and carbide distribution appeared similar to that in corresponding material produced by hammering, cogging, and rolling.

CONCLUSIONS

Work described here has shown that the operation of a plant to the general design described can be undertaken on a works scale to produce continuously cast billets of 4-in. square section of alloy steel. There are, however, limitations regarding the speeds which may be used, depending on the particular steel cast. For successful continuous casting, particular attention must be paid to steelmaking and ladle practice, and tighter control is needed than can be tolerated in standard ingot practice.

The structure of the so-called 'billet' produced by continuous casting was essentially similar to that of an ingot of comparable section. Normally the structure of the continuously cast billets produced during this work was slightly inferior to similar ingot structures, but under certain casting conditions it was slightly better. In no cases was a marked improvement obtained. In particular, 18–4–1 steel cast under conditions which gave satisfactory plant operation on one strand in accordance with the plant design did not give a satisfactory billet structure, because of centreline shrinkage, porosity, and the occurrence of carbide segregates. Water-spraying and electromagnetic stirring gave no significant improvement in the structure of 18–4–1 high-speed steel.

The economics of the process have not been considered in this paper, but investigation showed that the casting of 18–4–1 high-speed steel would be an economical proposition at 36 in./min casting speed if the resultant bars were acceptable in quality. However, it has been shown that satisfactory quality can only be obtained by casting at below 15 in./min which causes handling and casting difficulties.

To cast 18–4–1 high-speed steel successfully it may be necessary to alter the method as follows:

(1) Alteration of the section cast, e.g. to oval or rectangular section
(2) Modification of the pouring system to permit lower casting speeds
(3) Modification of the stirring method to avoid carbide segregation.

Stainless steel (18–8) can be produced through the continuous-casting machine with satisfactory quality, but investigation of the economics of this particular plant suggested that the increase in yield obtained was insufficient to balance the extra expense compared with conventional casting techniques.

Acknowledgments

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Hydrogen in Ferrosilicon

ANALYTICAL METHODS AND AVERAGE HYDROGEN CONTENTS OF SOME COMMERCIAL GRADES

By N. Christensen and K. Gjermundsen

SYNOPSIS

Apparatus and methods have been developed for a determination of the total hydrogen content of ferrosilicon, with a view to avoiding errors caused by volatile impurities. Vacuum-fusion of previously vacuum-heated samples indicates the presence of residual hydrogen.

Hydrogen in ferrosilicon shows severe segregation depending on the presence of hydrated lime and alumina in the sample. The hydrogen content increases with decreasing particle size, increasing atmospheric exposure, and rapidity of analysis after manufacture.

1519

INTRODUCTION

IN RECOGNITION OF the adverse effects of hydrogen, steelmakers are attempting to control the amount introduced into the steel after the period of boiling. A rough estimate indicates that the deoxidizers are probably not a major source of hydrogen. In mild and low-alloy steel grades any potential hydrogen originally present in the ferroalloy additions will usually be reduced by a factor of about 50 to 100 when converted to the basis of 100 g of steel. It would appear, therefore, that contributions from the deoxidizers might be of little importance except in cases of carefully controlled steels; nevertheless, it seems worthwhile to establish the ranges of hydrogen contents to be expected in some of the more commonly used grades of ferrosilicon.

A summary of published analytical data for commercial grades has been made in Table I. Some of these figures, e.g. those reported by Hatfield and Newell,[1] are average values obtained over a period of several years. The Russian authors Eljutin, Pawlow, and Lewin[3] state that contamination of the steel through ferrosilicon addition may at times occur.

It will be noted that the figures reported by Hochmann[6] are the only ones measured by vacuum-heating. The nitrogen content determined in this way was too low as compared with wet chemical methods; the amount of hydrogen found by hot extraction could not be controlled because evaporation interfered in the vacuum-fusion method.

Liang, Bever, and Floe[7] measured solubilities at 1400° and 1600° C in the range 5–25 p.p.m. It is not known to what extent equilibrium will be approached under conditions prevailing in pouring from temperatures well above 1600° C.

APPARATUS AND ANALYTICAL PROCEDURE

Two pieces of apparatus of conventional design have been used: one for vacuum-heating analysis in the temperature range 650–900° C, and one for vacuum-fusion. Both pieces of apparatus are equipped for a complete chemical analysis of the extracted gases; however, for reasons to be discussed later, only H_2O, H_2, and the total gas contents have been reported.

The vacuum-fusion apparatus is operated at about 1500° C and a nominal speed of pumping of about 12 l./s. Under these conditions a serious loss of hydrogen is encountered with alloys containing significant amounts of calcium, probably through the formation of calcium hydride in the film of condensed metal. An arrangement has therefore been provided for external heating of the quartz-glass furnace tube to about 650° C after each fusion.

On the average, three to five samples may be tested in one crucible. Alloys containing more than about ½%Ca require a more frequent replacement, a double or even a triple desorption being necessary in these cases. This eventually causes the metallic deposit to distill into the quartz-to-glass junction, when the furnace section must be disassembled and cleaned.

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Various methods have been tried to overcome this and other analytical difficulties. A brief presentation of these experiments will be made in a separate section at the end of the paper.

EXTRACTIBLE AND RESIDUAL HYDROGEN

In view of the simplicity of the vacuum-heating method, which appears to be widely used in the U.K.,[8, 9] it was thought desirable to test the completeness of extraction in vacuum-heating. The residual hydrogen content was determined by vacuum-fusion of samples that had already been analysed by vacuum-heating at 650° or 900° C. These samples were partly in the form of 3·5-g pieces, and partly crushed and screened to a particle size between 0·8 and 1 mm diameter. The latter ones were contained in previously degassed mild-steel capsules about 1 g in weight. In all cases great care was taken to avoid those parts of the original lumps that had been exposed to the atmosphere.

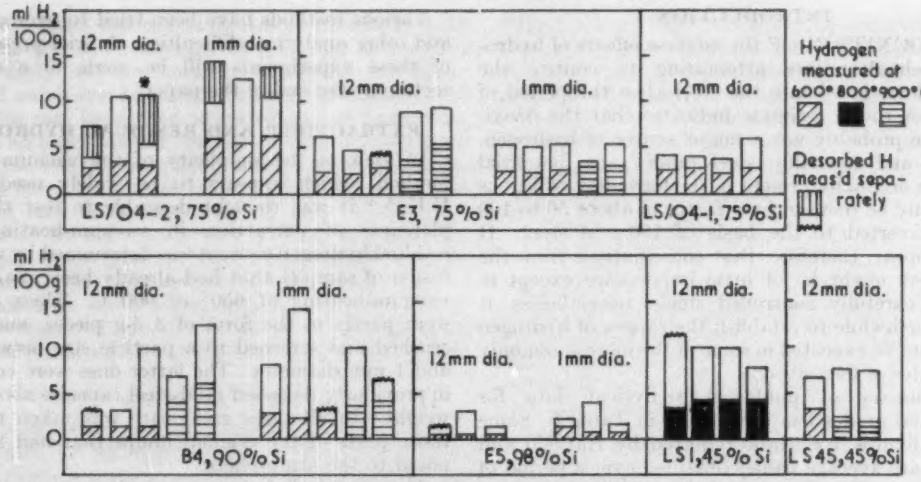
The results of these tests are shown diagrammatically in Fig. 1. Disregarding for the moment the variations in the absolute values of the measured hydrogen contents, and postponing the discussion of other details of the histograms in Fig. 1, it may be stated that a significant fraction of hydrogen has been recovered in vacuum-fusion of previously vacuum-heated samples. This is true also when the vacuum-heating was performed at 900° C, a conclusion which has been confirmed later by fusion of two control samples analysed at 900° C in the B.I.S.R.A. laboratories at Sheffield. An independent control of the reproducibility in vacuum-heating at 650° to 700° C has been obtained through cooperation with Metallografiska Institutet in Stockholm: at 700° C, a crushed and screened sample (75%Si) was found to contain $4·2 \pm 0·1$ ml H_2/100 g, whereas vacuum-heating at 650° C in the authors' laboratory showed $4·1 \pm 0·3$ ml/100 g.

Whilst the presence of residual hydrogen even after vacuum-heating at 900° C appears reasonably well established qualitatively, it is not possible to make quantitative statements on the basis of the available data. The fraction not recovered in vacuum-heating seems to depend, as much on the kind of alloy and on the particle size, as on the extraction temperature. This is also true for the composition of the extracted gases, as can be seen from Table II. The fraction of hydrogen in these gases varies between 48 and 96%, and it may be quite misleading to report the total dry gas as hydrogen. The presence of moisture will be considered in a separate section.

SEGREGATION OF HYDROGEN

The scattering of results to be seen in Fig. 1 might conceivably originate from some unknown error in the analytical procedure. Some alloys, on the other hand, have yielded fairly consistent values when tested over a period of several years. This points to heterogeneity as a more probable source of scattering.

An inspection of Fig. 1 reveals that the scattering of results is not eliminated by a sub-division of the alloy to a particle size below 1 mm diameter. Additional data not shown here confirm the trend indicated by Fig. 1: the use of a crushed and screened sample



* The LS/04–2 sample with asterisk has been fused after vacuum-heating without breaking the vacuum

Fig. 1—Examples of vacuum-fusion of previously vacuum-heated samples

reduces the scattering with alloys tending to give consistent results when tested in lump size, but does not improve the reproducibility much with other alloys.

Taking as an example alloy No. B4 containing 90% Si, it is seen that two determinations of residual hydrogen have given values of 5 and 12 ml H_2/100 g. These data were obtained with samples containing about 3000 particles each. This difference of about 7 ml/100 g must be accounted for by excessively high hydrogen contents of a few particles, say 5 or 10, since it is impossible for any large number of 'abnormal' particles to accumulate in one of the two halves of a thoroughly mixed sample. To this order of magnitude then, the residual hydrogen content of such particles would be

$$\left(7\ \frac{\text{ml}}{100\ \text{g}}\right)\frac{(3000\ \text{particles})}{(5\ \text{to}\ 10\ \text{particles})} \sim 3000\ \text{ml/100 g}.$$

A concentration of this magnitude strongly indicates the presence of hydrated lime or alumina, which in turn could be derived from the respective phosphides or carbides through reaction with atmospheric moisture.

Whilst a quantitative verification of this assumption cannot be given at present, there are qualitative indications of a segregation of calcium. This has been demonstrated with 1-mm particles of cast No. LS/04–2 by dropping one particle after another into a thin layer of a gelatine solution to which phenolphthalein

Table I

PUBLISHED DATA ON GASES IN FERROSILICON

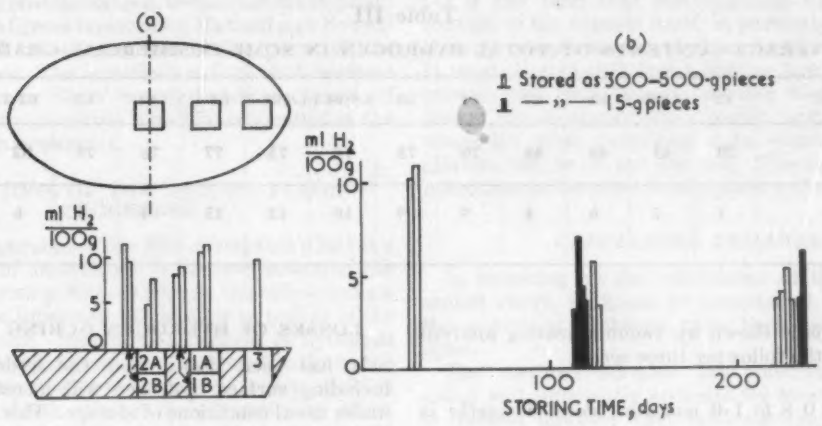
Grade, (% Si)	Hydrogen, ml/100 g	Oxygen, p.p.m.	Nitrogen, p.p.m.	Reference	Method
20–30	2	20	20	(1)	Vacuum-fusion
40–50	8–9	30–170	10	(1)	Vacuum-fusion
	> 3	...	...	(2)	Not known
	8	...	...	(3)	,, ,,
	3*	50*	10*	(4)	Vacuum-fusion
70–78	6–9	360	10–50	(1)	Vacuum-fusion
	7	...	...	(3)	Not known
	30	240	30	(4)	Vacuum-fusion
	11	190	traces	(5)	,,
	13	(14)	> 40	(6)	Extraction at 1200° C
78–85	12–42	10–140	5	(1)	Vacuum-fusion
90	3–6	...	...	(2)	Not known
	28	...	...	(3)	,, ,,

* Grade of alloy not stated

Table II

EXAMPLES OF AVERAGE CHEMICAL COMPOSITION OF EXTRACTED GASES

Grade	Particle Size, mm	Temperature, °C	Extracted Gas, ml/100 g			H_2, % of Dry Gas
			H_2	CO + N_2	H_2O	
45% Si (LS–45–3)	10–12	650	3·2	1·0	1·9	76
	0·8–1·0	650	3·1	1·6	1·7	66
	10–12	900	2·1	2·3	1·3	48
	0·8–1·0	900	3·9	1·7	1·4	70
75% Si (E3)	10–12	650	2·4	0·1	0·2	96
	0·8–1·0	650	2·4	0·6	0·1	87
	10–12	900	4·6	0·6	0·5	88
	0·8–1·0	900	3·0	0·6	0·2	83
90% Si (B4)	10–12	650	1·2	0·4	0·2	75
	0·8–1·0	650	2·3	1·1	0·1	68
	10–12	900	5·0	0·8	0·1	86
	0·8–1·0	900	3·8	0·9	0·2	81
98% Si (E5)	10–12	650	0·2	0·1	0	67
	0·8–1·0	650	0·8	0·8	0·3	50



(a) Spatial distribution of hydrogen (b) Effects of storage

Fig. 2—Measurements of sand-cast 75% ferrosilicon

had been added as an indicator. A distinctly pink halo indicating the presence of lime was observed around a few particles out of a total of one thousand. The exact number of such particles cannot be stated precisely, for this number will depend on an arbitrarily chosen limit of colour intensity.

Another indication of short-range heterogeneity has been found in alloy LS/04–2 through measurements performed on a particle fraction below 300 mesh. This fraction was obtained as a by-product in the preparation of samples of nut size, and it is therefore not a result of a prolonged process of crushing. A comparison of the contents of calcium, aluminium, phosphorus, carbon, and hydrogen in this fraction with the corresponding figures for nut-size samples shows that these elements have accumulated in the fines, as would be expected on account of the brittleness of the carbides and phosphides:

Particle Size	Wt-%				Gas Content, ml/100 g		
	Ca	C	Al	P	H_2O (650°C)	H_2 (650°C)	H_2 (residual)
10—12 mm	0·7	trace	2·5	0·02	0·1*	3	6
< 300 mesh	1·8	0·09	3·9	0·06	12·5†	185	22

* Freshly prepared samples † Samples stored in desiccator

Further observations in support of the postulated reaction with atmospheric moisture will be presented in a subsequent section dealing with atmospheric exposure. Although the non-uniform distribution of potential hydrogen is not a result of true primary micro-segregation, it has the same effect and must be considered in the appraisal of analytical data.

In contrast to the apparent microscopic segregation, a true segregation on a macroscopic scale appears to be less pronounced. An indication of the distribution of hydrogen in sand-cast ferrosilicon (75%Si) is given in Fig. 2. By comparison of the total hydrogen contents measured over a period of 25 to 45 days after pouring, it is seen that the hydrogen content of the upper central parts tends to be slightly lower than the average content.

EFFECTS OF ATMOSPHERIC EXPOSURE

The hydrogen contents presented hitherto have been measured, either after removal of the original surface of the sample, or with samples that have been carefully protected against atmospheric humidity after crushing. In the case of powder < 300 mesh reported above, the precautions have not been sufficient to prevent a reaction with water vapour during the preparation of the sample. A less rapid pick-up of moisture is observed with samples of a larger particle size as well; however, the state of this surface hydrogen appears to depend as much on the properties of the alloy as on the particle size.

A nearly reversible adsorption of water was shown by a 75% silicon grade LS/04–1, which had been stored as pieces of walnut size for several months at 20° C and 60% relative humidity. At the end of this period the average bulk content was 4·8 ml H_2/100 g, as measured in the central parts of several lumps, whilst a test including the exposed surface gave the following result:

Vacuum-heating at 650° C	8·4 ml H_2O/100 g 3·0 ml H_2/100 g
Subsequent fusion	2·1 ml H_2/100 g
Total, incl. surface	8·4 ml H_2O/100 g 5·1 ml H_2/100 g

Similar results were obtained with freshly crushed and screened samples of the same alloy (particle size 0·8–1·0 mm dia.). When tested at 650° C after various periods of exposure, the gas content including surface hydrogen was found virtually constant with one exception:

Exposure, h	0	13	18	36
H_2O, ml/100 g	0·7	5·4	1·1	1·0
H_2, ml/100 g	2·3	2·8	2·1	2·1
CO, etc., ml/100 g	2·8	5·0	2·5	2·6

In other alloys, exemplified by grade No. LS/04–2 (75% Si), the adsorbed moisture could not be removed

Table III

AVERAGE CONTENTS OT TOTAL HYDROGEN IN SOME COMMERCIAL GRADES

Grade No.	T2	LS45	E1	E2	P1	B1	LS/04-1	LS/04-2	B2·1	E4	E3	B2·2	B4	E5
Silicon, %	29	45	48	49	70	75	75	75	77	78	79	82	90	98
Hydrogen, ml/100 g	1	7	6	4	9	9	10	12	15	4	6	6	9	3

entirely. This is shown by vacuum-heating analysis at 650° C of the following three series:

(*a*) *Particles* 0·8 *to* 1·0 *mm dia.; stored* 4 *months in desiccator*

Exposure, h	0	6	24
H_2O, ml/100 g	2·6 and 1·3	1·3 and 1·0	4·0 and 2·6
Dry Gas, ml/100 g	7·6 and 6·5	6·6 and 7·7	10·7 and 8·9

(*b*) *Particles* 0·4 *to* 0·8 *mm dia.; degassed at* 800° *C prior to exposure*

Exposure, h	0	6	12	24	48
H_2O, ml/100 g	0	6·5	7·8	10·6	11·6
H_2, ml/100 g	0	2·6	2·5	2·3	0·5
Other Gases, ml/100g	0	4·4	6·8	10·9	11·2

(*c*) *Powder* < 300 *mesh* (same material as described on p. 251)*; degassed prior to exposure; saturated air*

Exposure, h	0	2	4	6
H_2O, ml/100 g	0	1·8	0·8	4·3
Non-condensible, ml/100 g	0	29·0	33·5	36·6
Condensible, ml/100 g	0	38·1	36·3	32·3

In series (*a*), the content of dry gas is seen to increase with time without a corresponding increase of desorbed moisture. Series (*b*) shows increasing amounts of moisture with increasing exposure time; in contrast to grade No. LS/04–1, however, the present alloy also gives off increasing amounts of other gases. Series (*c*) indicates a chemical reaction even more strongly. In this case the hydrogen content could not be measured because the gases derived from the sample 'poisoned' the palladium tube in a short time. After drying over P_2O_5 a condensible fraction has been determined by successive compression in the McLeod gauge. The saturation pressure of this component is about 20 mm Hg at room temperature. It appears to decompose slowly into constituents that are not condensible. These observations are in reasonable agreement with published data for hydrogen hemiphosphide P_2H_4, which has a boiling point of 57° C and which decomposes to PH_3 and $P_{12}H_6$ through a photochemical reaction.[10] It appears thus that the different behaviour towards atmospheric exposure shown by these two alloys depends on a chemical reaction in one case, and on physical adsorption in the other. This difference may be of significance both with regard to shipping and storing of the alloy, and to the effect of redrying prior to use.

LOSSES OF HYDROGEN DURING STORAGE

It has been shown that the hydrogen content including surface hydrogen will increase with time under usual conditions of storage. This is not true for the concentration prevailing in the interior of the particles, as can be seen from Fig. 2. In spite of considerable scattering, a trend of lower hydrogen contents with increasing time can be discerned. This effect appears to be more pronounced during the first four months; after this period there is hardly any change to be noticed. This process of diffusion occurs simultaneously with any surface reaction. Thus the filled columns in Fig. 2 indicate an average bulk content after 4 months slightly higher than 5 ml H_2/100 g, and after another 4 months slightly lower than 5 ml H_2/100 g. One of these pieces, when analysed with its original surface intact, gave a total content of 13·5 ml ($H_2 + H_2O$)/100 g.

A reduction of the bulk hydrogen content has also been observed with other alloys, e.g. grade LS/04–2 shown in Fig. 1. When tested 11 days after pouring (particle size 0·8–1·0 mm; columns to the right) the extractable hydrogen content was nearly twice that obtained after 45 days (particle size 10–12 mm; columns to the left). As a further example, a drop of the bulk hydrogen concentration from about 8 to about 3 ml/100 g has been observed in a 45% silicon grade after 3 years' storage.

Although the effects of storing observed so far are not sufficient for any quantitative evaluation, they indicate a clear trend of loss of hydrogen during the early stages. This conclusion has been confirmed and amplified by measurements by S. Velken of the ELKEM laboratories in Kristiansand.[11] When tested a few hours after pouring, the content of diffusible hydrogen in 75% silicon grades is higher than in stored material by a factor between 2 and 2·5.

HYDROGEN CONTENTS OF SOME COMMERCIAL GRADES

All of the present work has been carried out with samples taken from the current production of Norwegian works. Although the data collected so far are by no means complete, it is believed that they give a reasonably representative indication of the average bulk hydrogen contents to be expected with the present methods of production.

The results are shown in Table III. By comparison with the published data in Table I, it is seen that the latter ones are largely higher. This is believed to

depend mainly on differences in techniques of sampling; at any rate the figures reported by Hatfield and Newell[1] were obtained on samples containing parts of the exposed surface. The contribution from such surfaces may have more than counterbalanced losses of hydrogen, if any, resulting from film adsorption in the vacuum-fusion apparatus.

LIMITATIONS OF THE VACUUM-FUSION PROCEDURE

A correct appraisal of the film-adsorption effect is a major source of uncertainty in the comparison of the present data with published figures, this effect being a function of the properties of the alloy as well as of the apparatus design. Selective adsorption of hydrogen has been predicted,[12] and was confirmed early in the present investigation when only one pump of 2 l./s capacity was used. Under those conditions about 75% of the hydrogen released from the molten metal was lost by adsorption, even with alloys containing no more than 0·17%Ca. With two pumps in series, giving a nominal capacity of 12 l./s, the adsorbed fraction may be reduced to a range between 15 and 60% with such alloys, the high figures applying to tests performed with a deposit already present from the fusion of previous samples.

Alloys containing 0·7%Ca, as grade No. LS/04–2, may produce enough deposit during the first or second test to give an adsorption close to 90% at a pumping speed of 12 l./s. The following data, obtained with alloy LS/04–2 at a nominal speed of pumping about 70 l./s, indicate that significant errors may arise even with the first sample when the calcium content is high:

Collected gas	H_2, ml/100 g	$(CO + N_2)$, ml/100 g
Fusion	3·15*	6·40
Desorption	2·85	0·32
Total	6·0	6·7

* Including 0·5 ml H_2O/100g

Preliminary experiments using a bath of about 25% Si in iron for dilution of the sample have not been successful so far. This arrangement has proved satisfactory with the first two or three samples of low-calcium ferrosilicon, but even in this case a gradual accumulation of condensed metal has been observed. It remains to be seen whether a combination of dilution and of increased speed of pumping will prove adequate in preventing losses of hydrogen.

A dilution of 75%Si and higher-grade samples will probably also be beneficial in preventing a conversion of the graphite crucible into one of silicon carbide. The indications of chemical reaction of this kind are not only a change of the graphite lustre after fusion of the first sample, but also a rapid penetration of the walls, metal sputter being frequently observed at the inside of the radiation shield. After fusion of about 15 g of high-grade ferrosilicon there is usually no button of metal left in the crucible.

As a consequence of the formation of silicon carbide, the reduction of oxides is not complete with undiluted high-grade alloys. When such alloys are charged in a crushed and screened state in mild-steel capsules, it is frequently observed that the evolution of CO is less than that corresponding to the oxygen content of the capsule itself, in particular when more than one sample has already been fused in the crucible. It must be accepted that a similar lack of reduction prevails also with nut-size samples where this effect cannot be demonstrated directly, and it must be concluded that published data obtained without dilution will be on the low side. The nitrogen determinations on the other hand appear to be less affected.

CONCLUDING REMARKS

In summing up the conclusions of the work presented above, it should be emphasized that some of the analytical problems have been barely touched upon.

The methods adopted are considered basically sound and sufficiently accurate for most purposes; it is realized however that both equipment and procedures should be simplified for routine work. In such cases where a knowledge of the residual hydrogen content is required, it is hoped that further work along the lines of dilution and of high rates of pumping will prove fruitful. In particular, dilution appears to be a necessary step when figures for the oxygen content are required, as has been suggested more than 10 years ago for non-ferrous materials by H. A. Sloman.[13]

The scattering of results obtained with the vacuum-fusion method is believed to be an inherent characteristic of commercial ferrosilicon. It might prove feasible to reduce or avoid this scattering through a proper preparation of the sample, e.g. by a more extensive crushing and screening than applied in the present investigation. Such a procedure is open to criticism, in particular from a consumer's point of view, because the sample would not be representative in the absence of hydrated lime and alumina. A proper solution to this problem might consist either in accepting the scattering as a piece of information adding to a description of the material, or in the use of two separate figures, i.e. one for the hydrogen content of the matrix of the alloy and another giving the maximum hydrogen content corresponding to a complete transformation of carbides and phosphides into hydrated oxides.

A similar problem of sampling is encountered with regard to surface hydrogen. In this case it would seem reasonable to include the surface moisture in the total hydrogen content whenever a roasting treatment will not remove the moisture as such. From the preliminary work presented here on this question, it appears that the adsorption will be more or less irreversible depending on the presence of carbides and phosphides in the alloy. It must be concluded therefore that the problems of sampling cannot be properly settled until these effects have been more thoroughly studied.

Acknowledgments

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They also desire to thank institutions and individuals who have contributed with control determinations, namely Metallografiska Institutet, The British Iron and Steel Research Association, and Mr. S. Velken of Elektrokemisk A/S. Thanks are also due to staff members of SINTEF's metallurgical section, in particular to Mrs. E. J. Dörrum and Mr. T. Simonsen.

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Open-hearth Furnace Regenerators

A SURVEY OF THEORETICAL AND EXPERIMENTAL WORK WITH RECOMMENDATIONS FOR THE CALCULATION OF PREHEAT

By the B.I.S.R.A. Regenerator Group

PREFACE

IN 1956 the Steelmaking Instruments Sub-Committee examined much information on open-hearth regenerator practice, including the following factors:

(1) Air velocity through the checkers
(2) Air volume per unit surface of checkers
(3) Checker surface area per unit brick volume
(4) Checker surface area per unit air flow per hour
(5) Regenerator shape (height to length)

only to find that no helpful conclusion could be reached on how to increase the level of preheat temperature in an open-hearth furnace. Hence the Sub-Committee formed, in January, 1957, the Regenerator Group with the remit "to summarize the present knowledge of regenerator theory and practice, to indicate what further gains in furnace performance might be achieved by improvements in regenerator efficiency, and to recommend the course further investigations should take."

This Group, the constitution of which is given below, was able to make a quick start because two of its members, Dr. W. P. Cashmore and Mr. J. V. Edwards, immediately put at its disposal their own private surveys on the experimental (Part I) and theoretical (Part II) aspects of the subject respectively, and Mr. D. Meachen and Mr. K. W. Bell jointly drew up a recommended method for calculating preheat in a regenerator system (Part III). Although in the course of discussion during several meetings many changes were suggested and agreed upon, the Group, while accepting responsibility for the report, acknowledges that the work of these four members forms the basis of the report.

SYNOPSIS

This survey covers experimental work by many investigators into the various factors affecting regenerator design. It is shown that care and attention to design, sealing, and cleaning of the regenerator increase both the preheat level obtained and the working life of the regenerator, and thus increase the output and availability of the furnace.

A review is included of some experimental and theoretical work on the calculation of heat transfer in regenerators. Graphs are given to enable the relative effects of changing design to be quickly determined, in specific cases, without recourse to calculation.

A recommended method for calculating heat transfer in a regenerator is demonstrated by a worked example. The calculation is quick and easy and, for a typical regenerator, takes about 30 min.

1603

Paper SM/AH/103/58 of the Steelmaking Instruments Sub-Committee of the Steelmaking Division of the British Iron and Steel Research Association, received on 4th July, 1958.

The constitution of the Regenerator Group was as follows: Mr. W. C. Heselwood (*Chairman*), United Steel Companies Ltd.; Mr. K. W. Bell, B.I.S.R.A.; Mr. A. J. Brown, B.I.S.R.A.; Dr. W. P. Cashmore, formerly of Guest Keen Iron and Steel Co., Ltd., now with Algoma Steel Corporation, Canada; Mr. C. W. J. Crawford, B.I.S.R.A.; Mr. J. V. Edwards, Steel Company of Wales Ltd.; Dr. C. Hulse, Sheffield University; Dr. G. R. Mattocks, British Glass Industries Research Association; Mr. R. Mayorcas, John Summers and Sons Ltd.; Mr. D. Meachen, B.I.S.R.A.; Dr. J. Pearson, B.I.S.R.A.; Mr. D. Thomas, Park Gate Iron and Steel Co., Ltd.; Mr. P. Start, Sheffield University; and Mr. J. W. Till, John Summers and Sons Ltd.

LIST OF SYMBOLS

A	Total exposed heating surface of checker bricks, ft²
B	Atmospheric pressure, lb/in²
C	A constant
c_p	Specific heat of a gas at constant pressure, B.t.u./lb/° F
d	Duct width, in.
h_r	Total frictional pressure loss over regenerator height, in. w.g.
L	Total regenerator height, ft
m	A factor
n	A factor
r	Standard heat resistance
$T°\,K = °C+273$	Absolute gas temperature (Kelvin)
W	True velocity of gas, ft/s
W_0	Velocity of gas corrected to N.T.P., ft/s
a_c	Convective heat-transfer coefficient for waste gas or air, B.t.u./ft²-h-° F
$a\,(co_2+H_2O)$	Total radiation heat-transfer coefficient from water vapour and CO_2 for a particular case, B.t.u./ft²-h-°F⁴
$a_{corr.}$	Correction for true temperature conditions to be added to $a(co_2+H_2O)$, B.t.u./ft²-h-° F⁴
a_{GR}	Total radiation heat-transfer coefficient waste gas to brick, B.t.u./ft²-h-°F⁴
γ	True density of a gas, lb/ft³
γ_0	Density of gas at N.T.P., lb/ft³
λ_g	Thermal conductivity of a gas, B.t.u./ft²/h/° F
λ	Excess air factor.

Part I—A Critical Survey of Experimental Work on Open-hearth Regenerators

INTRODUCTION

In this part of the report it is intended to survey the existing literature on the measurement of preheat and its influence on open-hearth furnace operation.

The feature which distinguished the open-hearth furnace from its predecessors was its regenerators, the introduction of which was necessary before Siemens was able to make steel using the relatively low-grade fuels at his disposal. It is therefore not surprising to find that the temperature of the incoming air is a critical factor affecting furnace operation. Although many early steelmakers were aware of this fact, a considerable time elapsed until instruments could be developed which would enable the air temperature to be measured with any degree of accuracy.

Bansen[1] was well aware of the intimate relationship which existed between preheat and furnace performance but, owing to the limitations of the instruments available to him, he was forced to use the temperature of the top courses of the checkers as an estimate of the preheat. He was able to show that both working rate and fuel consumption were considerably affected by variations of this quantity (*see* Fig. 1). Other notable workers in this field were Schack,[2] Trinius,[3] Finlayson and Taylor,[4] and Heiligenstaedt.[5] Schack used a type of suction pyrometer which he had developed to measure air and waste-gas temperatures at various points in the furnace. After this pioneer work little seems to have been done until Marsh[6] related furnace performance to preheat temperature at Bethlehem Steel Corporation. The next measurements were made by Larsen[7] at the United States Steel Corporation, while subsequent work has been

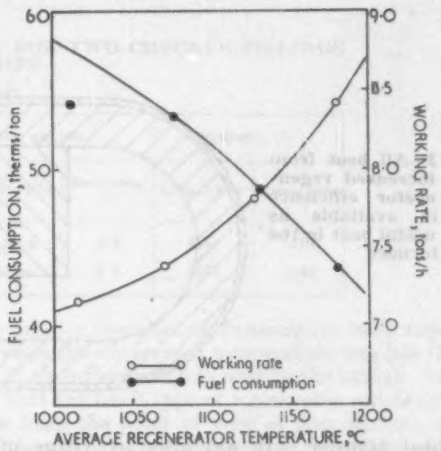


Fig. 1—Relationship between regenerator temperature, working rate, and fuel consumption (Bansen[1])

done by Cashmore[8] and also by Carney, Oravec, and van Meter.[9] The results obtained by all workers show the beneficial effect of preheat on open-hearth performance. They will be discussed in more detail later. Recently some measurements of waste-gas temperatures have been carried out by Kornfeld, Voitov, and Koshelev[10] (*see later*).

It should be realized that procedures for measuring preheat vary widely and comparison of various investigators' results can, therefore, be misleading. To enable comparisons to be of value the Regenerator Group adopted the following definition of preheat temperature for its investigations:

> Preheat temperature is the temperature of the combustion air during the refining period at the moment half way through the air cycle, at a position not less than 12 in. from the uptake wall and between stage level and the point where the uptake turns into the air port.

GENERAL CONSIDERATIONS

Before discussing the experiments referred to in greater detail, it seems worthwhile to consider the general background against which these experiments must be interpreted.

It is evident that in the open-hearth furnace, however high the rate of heat transfer, the waste gases can never leave the furnace laboratory at a temperature lower than that of the surface of the slag. (In practice this statement requires some modification to allow for the effect of heat losses and infiltrating cold air, but the general principle remains the same.) This means that the combusting gases can give up only a relatively small fraction of their sensible heat before leaving the furnace laboratory and, unless some steps are taken to recycle some of this, the heat requirement per ton of steel is extremely high. Thus the regenerative principle becomes more important the lower the grade of the fuel and the higher the charge surface temperature. Indeed, for certain low-grade fuels it would be impossible to make steel without regenerators. Thus the regenerative effect increases the fraction of the heat in the flame available

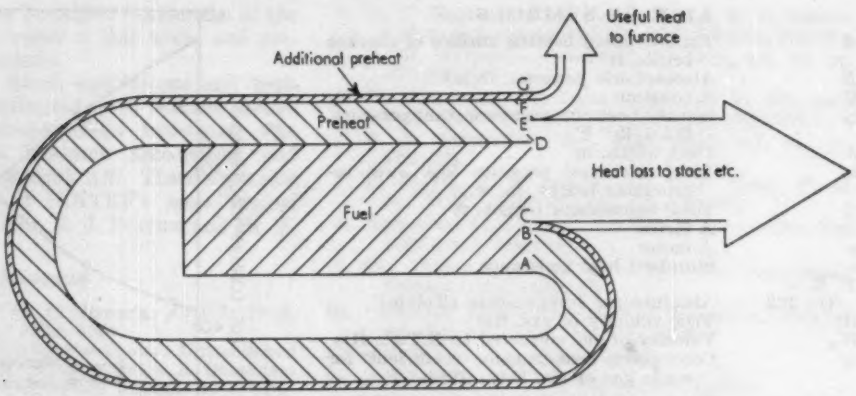


Fig. 2—All heat from increased regenerator efficiency is available as useful heat in the furnace

for heat transfer. It will also, by virtue of the increased flame temperature, be possible to enhance the heat-transfer rate itself.

It is evident that what is of real interest, strictly speaking, is not so much the preheat level itself as its relation to the waste-gas temperature, since the latter gives a measure, for given combustion conditions, of the rejected heat, and the preheat in relation to this temperature provides an indication of the proportion of this heat which is recycled into the furnace. However, some investigators have reported that the waste-gas temperature is surprisingly constant within the range 1600–1650° C *during the refining period*. Although the absolute accuracy of waste-gas temperature measurements made hitherto has been questioned, the readings appear to be consistent; it has been noticed at the Guest Keen Iron and Steel Co., Ltd., Cardiff, for instance, that the waste-gas temperatures do not vary significantly over a wide range of combustion conditions. This surprising conclusion has also been confirmed by Kornfeld, Voitov, and Koshelev[10] in the U.S.S.R., who found that there was no real change when up to 30% of oxygen enrichment was used in the combustion air.

It should be borne in mind, as Schack[11] has pointed out, that every unit of heat supplied as preheat is available for heat transfer, whereas only about a quarter of that supplied in the fuel eventually finds its way into the bath. Therefore, a false impression of the value of the heat recycled as preheated air may be given by assessing it in terms of the actual proportion of heat.

The effect of additional preheat on the useful heat, i.e. the amount of heat available before the gases are cooled to the temperature of the steel, is illustrated in Fig. 2. The quantity of heat in the gases leaving the furnace at 1600° C depends on the fuel and excess air but is independent of the preheat. If the amount of heat recycled as preheat is increased from AB to AC, the total heat is increased from AF to AG and the additional useful heat recycled is in the proportion $FG : EF$ rather than $FG : AD$.

It is possible that, as the standard of maintenance improves through careful sealing and continuous waste-gas analysis, the desirability of measuring waste-gas temperatures will become more apparent. It would seem that, having concentrated on the temperature of the air itself, we should be well advised to consider the waste-gas temperature.

Summing up, therefore, it is evident that the effect of preheat is to increase the heat release in the furnace laboratory and so improve the furnace performance. It must be remembered that as the campaign proceeds and changes occur in the regenerator system there will be a change in preheat. There will also be a considerable increase in the heat losses from the furnace laboratory, and these two factors between them will tend to reduce that fraction of the heat released which is taken up by the bath.

DISCUSSION OF RESULTS

The results of investigations can be divided under two headings:

(1) The effect of preheat on output rate and fuel consumption for a given regenerator design.
(2) The effect of a change in regenerator design on preheat, output rate, and fuel consumption.

Tables I and II show the effect of these features on working rate and fuel per ton. Table I shows that

Table I
EFFECT OF PREHEAT ON WORKING RATE AND FUEL CONSUMPTION

Origin	Furnace Size, tons	Increase in Working Rate per 100° C Rise in Preheat, %	Decrease in Fuel Consumption per ton per 100° C Rise in Preheat, %
Bansen		9	15
Marsh: Bethlehem Steel Works	179	13	Not quoted
Cashmore: Guest Keen Iron and Steel Co., Ltd., Cardiff	100	5	8
Carney, Oravec, and van Meter: United States Steel Corporation	116	7	Not quoted
Strähuber: Hückingen*	147	6·5 8·3	6·5 (Air) 8·3 (Gas)
Engels and Prieur	70	8–10	Nil

* These results are quoted as a 6·5% change in heat consumption for a 100° C change in air preheat; an 8·3% change in heat consumption for a 100° C change in gas preheat; and corresponding output changes

Table II

WORKING RATE, FUEL RATE, AND PREHEAT FOR TWO CHECKER FILLINGS AT G.K.I.S., CARDIFF

250-ton tilting furnaces

Furnace	Tons/week			Fuel Rate, gal/ton			Preheat, °C	
	12 in. × 9 in.	6 in. × 6 in.	Increase, %	12 in. × 9 in.	6 in. × 6 in.	Decrease, %	12 in. × 9 in.	6 in. × 6 in.
F	2350	2622	11·6	31·1	28·4	8·7	1043	1178
G	2289	2580	12·7	31·7	29·7	6·3	1077	1144

100° C preheat leads to 5–13% improvement in working rate and 8–15% reduction in fuel consumption.

The evidence would therefore appear to be that 100° C increase in preheat, if the average fuel input remains constant, will result in 5–10% increase in working rate and a reduction in fuel consumption of the same order. In the present state of ignorance there seems little point in trying to obtain a better estimate of the effect of preheat on working rate.

Strähuber[12] at Hückingen quotes the effect of 100° C change in both air and gas temperature on the performance of the Hückingen fixed furnaces, but he does not mention whether these results were based on experiments or calculations. The results, quoted in Table I, each refer to a particular regenerator under changing conditions and not to a change in regenerator design.

Some very interesting experiments were also carried out in Germany by Engels and Prieur,[13] in which supplementary heating of the checkers using coke-oven gas was investigated to increase furnace working-rate. The gas burned in the checkers varied from 16% to 19·7% of the total supplied, no extra air being introduced to burn this gas. On the basis of these experiments, it would seem that 100° C increase in preheat made a difference of 8–10% in the working-rate.

Measurements in the Regenerators

Many investigations have been carried out in connection with checker efficiency. The first of these would seem to have been carried out by Schleicher.[14] Gooding and Thring[15] also carried out an investigation on a model checker. The first investigation on an actual working checker seems to have been carried out by de Graaf and van Stein Callenfels,[16] who used a model checker to achieve a design, which was later built, in which flow and temperature were comparatively uniform across a horizontal section of the chamber. They eventually succeeded in this by building the top of the checker in steps and inserting, under it and transverse to the flow of gases, a series of walls increasing in height towards the slag pocket. This enabled the customary concentration of air towards the slag pocket and the waste gases towards the valve house to be avoided, but, as a result of calculation based on the measured temperature of the brickwork at several points in the regenerator, they came to the conclusion that the resulting uniformity had a negligible effect on the preheat level.

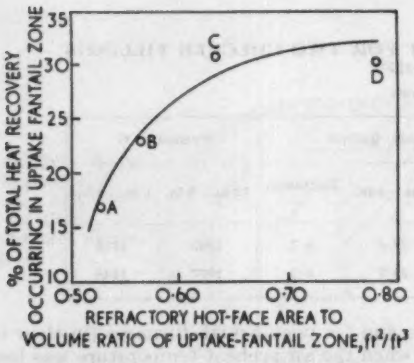
Baake and Stollberg[17] consider that a preheat of 1350° C for air and 1250° C for producer gas should be the aim for open-hearth furnaces (in their experience, when the air preheat temperature was less than 1200° C, air infiltration was usually the cause). They found that the usual type of regenerator construction is poor from the point of view of flow pattern, and hence the storage and regenerating capacities of the checkerwork are not used efficiently. To improve the distribution of gases over the regenerator cross-section, a checkerwork which causes the gas flow to change direction was used, i.e. a staggered basketwork, thus breaking up the flow and distributing the gases. They state that a frame brick introduced by Franz, Krug, and Ullsperger also gives a high preheat temperature because of improved gas distribution.

They concluded from model work, the appearance of the regenerator, the results of frame brick trials, and pressure, velocity, and temperature measurements above the checkerwork that the problem of gas distribution in the regenerator is of first importance and is more important than regenerator size.

This problem has also been considered from a theoretical point of view by Trier.[18] He had been able to show that non-uniformity in the air and waste-gas distribution which results in the air passing through one part of the checker and the waste-gases through another has a far worse effect on the preheat than is the case when the paths of the two coincide, even though this may lead to variation in the intensity of thermal loading of the checker surface. For example, Trier indicates that variations of 25% on either side of the means in gas and air velocity have a negligible effect on the preheat, whilst variations of 50% reduce it by about 15%. He also investigated a wide range of geometries of checker chambers, including horizontal and double-pass varieties, the suitability of the individual types being assessed on the basis of the following criteria:

(1) As much checkerwork as possible with little free space above and below the regenerator
(2) A long flow path for the gases inside the checkerwork (involving a longer period for retention of the gases in the regenerator)
(3) Adequate free space and with it a uniform distribution of flow
(4) Bends avoided as far as possible, for these induce pressure losses without corresponding heat exchanges.

These requirements are partially contradictory, and Trier gives examples of the best compromises for particular cases. He also considered the effect on the flow distribution of the buoyancy effect and shows that, when the pressure drop due to friction is of the same order as that due to buoyancy, heating the cold air as it passes upwards and cooling the gases as they



A Furnace *A*, all-basic *B* Furnace *B*, basic end
C Furnace *C*, extended hearth *D* Furnace *D*, standard silica

Fig. 3—Relationship between surface/volume ratio and heat-recovery efficiency in uptake-fantail zones of various open-hearth furnaces (Carney, Oravec, and van Meter[6])

descend tends to a greater uniformity of distribution.

Velocity and temperature measurements in the actual regenerator have been carried out by Marsh[19] and by Carney, Oravec, and van Meter.[6] Whilst their experiments are of great general interest their principal value would appear to be that they confirm the well known hypothesis that the incoming air passes through the checker at the pitside whilst outgoing waste gases tend to concentrate at the valve-house end of the checker. The distribution of the gases in a regenerator is sensitive to the inlet conditions, and simple modifications here which promote a smooth entry into the chamber may have a considerable effect on the flow. It does not seem that any attempt has been made to determine the effect of this non-uniformity on furnace performance.

Effect of Uptakes and Slag Pockets on Preheat

It is evident that the regenerative system of the open-hearth furnace consists not only of the checkers but also of the uptakes and slag pocket. Carney, Oravec, and van Meter have been able to show that the fraction of the preheat which is obtained by the additional regenerative effect occurring between the checkers and the top of the uptake can be related to the ratio between the heating surface of the slag pocket and uptake to the volume bounded by these surfaces (Fig. 3). It seems well worthwhile pursuing this idea; the only difficulty is to decide on the effective heating surface involved.

Cashmore[8] has suggested that the higher preheat observed on one of the two 100-ton furnaces he examined was due to the fact that it was a twin-uptake furnace and accordingly had a greater uptake area, which effectively increased the preheat.

Air Infiltration

The effect of air infiltration has been studied by Carney, Oravec, and van Meter. It was found that, on the first few heats after relining, the typical furnace has 25% air infiltration. After mechanically sealing the furnace ends and regenerators, and the natural sealing by fluxing of the refractory hot faces, the air infiltration decreases to 15–20% after 30–60 heats,

Table III
CHANGE OF PREHEAT WITH AGE

Furnace	Age, heats	Preheat, °C			
		Charging	Hot Metal	Melting	Refining
S.C.O.W. *N*	20– 27	1090	1100	1130	1060
	78– 85	965	1005	1045	1010
	133–140	870	820	968	954
S.C.O.W. *S*	20– 27	1130	1180	1245	1200
	78– 85	1084	1140	1200	1148
	133–140	1030	1085	1166	1100
G.K.I.S. *E* (Checkers blown weekly)	18– 94	1050		1110	1130
	114–158	1050		1090	1140
	179–235	1050		1100	1070
J. Summers *C*	36– 40	1180		1240	1255
	76– 80	1115		1200	1240

then gradually increases to 20–25% near the end of the campaign.

Haynes[20] found that, where air fans are used, infiltration to the ingoing air supply is limited to times when the supply of fan air is restricted, and also that under normal conditions, owing to the fact that the normal running range of positive air fans supplying steel-melting furnaces is below their full power, the actual air reaching the furnace port is much nearer the metered quantity than was hitherto appreciated.

Baake and Stollberg[17] found for a Maerz furnace that the excess air at the stack was 35% for a good furnace and 94% for a poor one, and this difference in air infiltration affected the preheat temperature by 250–300° C. Similar values for the excess air at the stack are reported by Till.[21] The excess air at the stack was measured on two furnaces and values of 86% and 77% were found. The furnace brickwork, particularly in the region of the regenerators, was examined and sealed where necessary. A second survey after sealing showed that infiltration had been reduced so that the excess air at the stack was now 42% for each furnace. It should be emphasized that the excess air at the stack includes that from all sources of infiltration in the flue system.

Effects of Deposits in the Regenerators

The cleaning of regenerators is a method by which a checker filling of smaller (and therefore more efficient) openings may be used with dirty gases

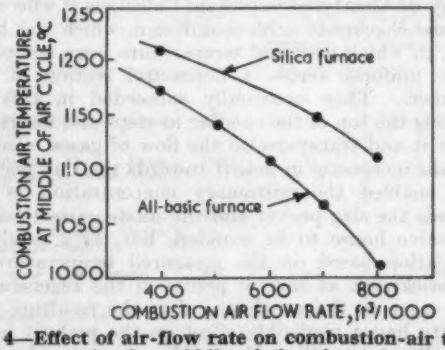


Fig. 4—Effect of air-flow rate on combustion-air temperature in the middle of the air cycle (Carney, Oravec, and van Meter[6])

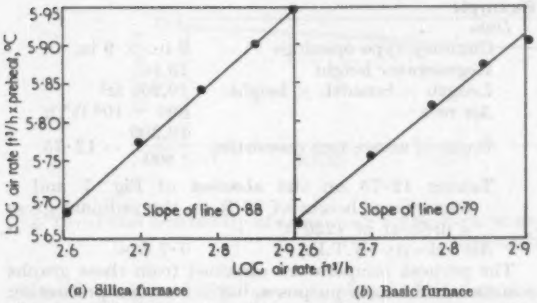


Fig. 5—Indication of variation of heat-transfer coefficient with air rate. Data for evaluation from Fig. 3

without losing its efficiency rapidly with age. In this connection, attention should be drawn to the ease of cleaning encountered with a horizontal regenerator. Three-pass horizontal regenerators were in use in open-hearth furnaces for 20 years without the bottom pass having to be renewed.[22] To clean regenerators, water, steam, compressed air, or mechanical means have been used. Washing out a hot regenerator with water has proved successful in many cases, and no damage to the checkerwork was noticed.[18]

The preheat from a regenerator is found to change with age, as shown in Table III. The Guest Keen Iron and Steel Co., Ltd. *E* furnace checkers, which are blown weekly, retain their original preheat throughout a campaign better than regenerators which are not cleaned.

There is no doubt that a higher preheat can be obtained from a given regenerator by using smaller openings (e.g. Table II). Normally, smaller openings cannot be tolerated because of the effect of the regenerator's age on preheat. If the regenerator is cleaned regularly the effect of age can be delayed, but it is not known how far the size of the openings can be reduced safely when a particular system of cleaning is used.

Effect of Air Rate on Heat-transfer Coefficients in the Regenerator

Carney, Oravec, and van Meter[9] were able to relate instantaneous preheats with air flow (Fig. 4). They appreciated that the values of preheat they obtained upon changing the air flow would not be the same as

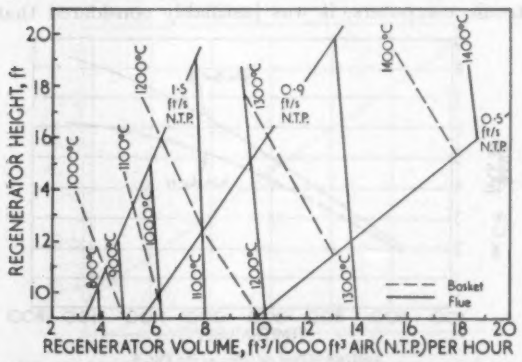


Fig. 6—Relationship of regenerator design and operating factors for 6 in. × 6 in. openings

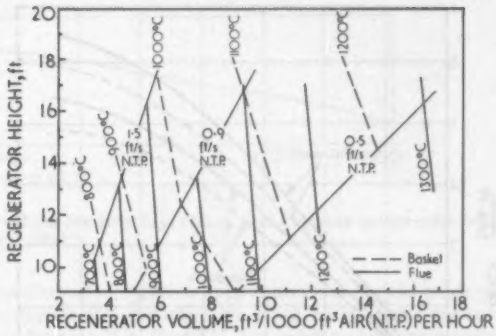


Fig. 7—Relationship of regenerator design and operating factors for 9 in. × 9 in. openings

those which would be obtained if the latter were maintained at its low values for any considerable period of time because the ratio of the heat capacity of air to waste gas would be different. In Fig. 5 the log (air rate × preheat) is plotted for two furnaces, one having a silica roof (two regenerator chambers) and the other of all-basic construction (one regenerator chamber). The overall heat transfer varies as the 0·8th and 0·9th power of the air rates. The difference in the exponents for these two furnaces is of a magnitude to be expected theoretically, from the difference in height of the regenerator fillings and the difference in regenerator volume.

Part III shows how the preheat to be expected from a proposed design may be determined, and from this point of view present knowledge of regenerator design is adequate. The validity of preheat calculations is discussed in Part II, where it is shown that methods for basketwork regenerators are more firmly established than for chimney regenerators. In the design of a regenerator for a particular practice, it is necessary to consider not only the preheat but also the regenerator life, the loss of efficiency with age, and the method of cleaning the regenerator. The relative effects of these factors are not known and so it is not possible to arrive at the best compromise when designing a regenerator.

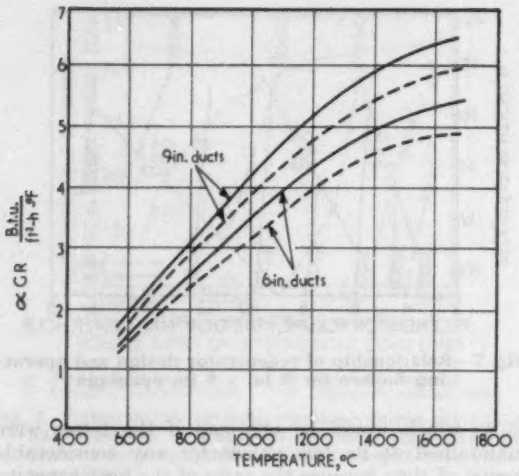
The complete design for a regenerator must depend on the use of the minimum brick opening permitted for the type of fuel used, the regenerator loading, and the effectiveness of routine regenerator cleaning. The relative effects of changing a regenerator design are shown by Figs. 6 and 7, the use of which is described in the Appendix to Part I, but it is not possible to choose the best design using these charts, since the use of various checker-cleaning methods and double-pass regenerators is not considered.

CONCLUSIONS

It has been shown that an increase of 100° C in the preheat level can lead to 5–10% increase in working rate and a corresponding reduction in fuel consumption if the average fuel input rate is kept constant. The value, therefore, becomes apparent of

(i) Waste-gas analysis as a check on air infiltration
(ii) Measurement of waste-gas temperatures in addition to preheat determination.

The investigations into checker conditions are



Analysis of fuel oil: C	86%
H_2	12%
S	2%
Analysis of tar fuels: C	89·5%
H_2	5·6%
S	0·43%
Mean temperature difference between gas and brickwork	200° C
Excess air	25% ($\lambda = 1{\cdot}25$)
Emissivity (refractory brick)	0·69

Fig. 8—Portion of radiation due to CO_2 and water vapour in waste gases from liquid fuels. (Note: The presence of water vapour arising from steam atomization of the fuel has a negligible effect on the calculated preheat)

discussed and it is suggested that, whilst considerable divergences from uniformity of flow distribution exist in the checkers, there has as yet been no real evidence that this has an appreciable effect on furnace performance, although this point is somewhat controversial.

The investigations of Carney, Oravec, and van Meter have shown the important part played by the furnace uptakes in contributing to preheat, and it is suggested that this aspect should be further investigated.

The dependence of the overall heat transfer upon the air rate can be determined by preheat readings taken immediately after several changes of air flow are made during one reversal. It would be particularly interesting to find how this relationship changes with the design and age of the regenerator.

APPENDIX TO PART I

The Determination of Preheat Temperature Using Figs. 6 and 7

Though the calculation of preheat for a particular regenerator is readily accomplished in a short time, it is sometimes useful to study whether a particular change in regenerator design will have a large or small effect. In most cases it is not necessary to make the calculation described in Part II, since an approximate result can be obtained from graphs based on these calculations.

These graphs can be used for any size of regenerator and any air flow rate per hour. The axes used are the height of the regenerator filling as ordinate and the volume of regenerator filling per 1000 ft^3/h of air (i.e. length × breadth × height per 10^3 ft^3 of air per h) as abscissa.

Example

Data

Chimney-type openings	9 in. × 9 in.
Regenerator height	15 ft
Length × breadth × height	10,200 ft^3
Air rate	800 × 10^3 ft^3/h
Ratio of above two quantities	$\frac{10{,}200}{800} = 12{\cdot}75$

Taking 12·75 on the abscissa of Fig. 7 and a regenerator height of 15 ft on the ordinate gives a preheat of 1220° C

Air velocity (N.T.P.) 0·7 ft/s.

The preheat temperature obtained from these graphs is sufficient for most purposes, but it is a wise precaution to check any result of particular interest, by calculation as described in Part III, before drawing a final conclusion.

Figures 6 and 7 should be used when it is required to find the effect of a change of air rate or regenerator size. For instance, if a change of regenerator height is proposed while keeping the size of regenerator opening and plan area the same, the effect is found by following a line of constant velocity. Three of these lines—0·5, 0·9, 1·5 ft/s N.T.P.—are shown. These lines intersect at a point, the origin, not shown in Figs. 6 and 7.

The validity of preheat calculations is discussed in Part II, where it is shown that the heat-transfer coefficient from air to brick for chimney-type regenerators is less well established than for the basketwork type. Thus, comparison of preheat values calculated for chimney-type and basketwork-type regenerators may be misleading.

The calculated preheat for 6 in. × 6 in. and 9 in. × 9 in. regenerators is shown in Figs. 6 and 7, when the regenerator dimensions and air rate per hour is known.

Assumptions made in calculating Figs. 6 and 7 were:

Brick thickness	3 in.
Waste-gas temperature inlet	1600° C

Distribution of air and waste-gas flow is uniform

Waste-gas contains 20% excess air

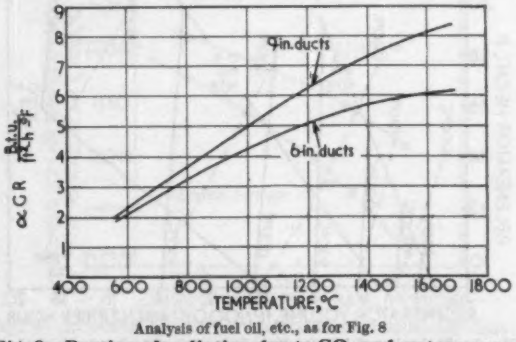
Mean waste-gas temperature for heat-transfer coefficient 1300° C

Mean air temp. for heat-transfer coefficient 550° C.

Part II—A Review of Theoretical Treatments of Regenerative Heat Transfer and Experimental Determination of the Heat-transfer Coefficients

METHODS OF CALCULATING HEAT TRANSFER IN REGENERATORS

The equations for heat transfer in a regenerator are complicated and, before the appearance of electronic computers, it was justifiably considered that



Analysis of fuel oil, etc., as for Fig. 8

Fig. 9—Portion of radiation due to CO_2 and water vapour in waste gases from coke-oven gas

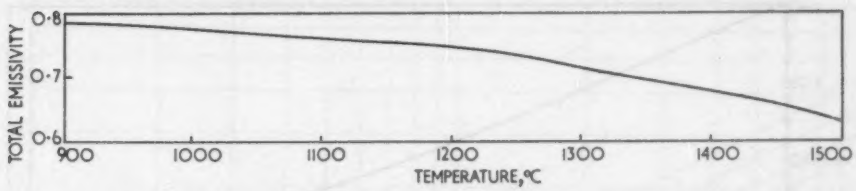


Fig. 10—Total emissivity of refractory brickwork (average values for fireclay, silica, and chrome magnesite bricks during furnace life[24])

even if a correct solution were obtained the results would be too complex to have any practical value. The result was that for many years regenerator calculations were based, not on mathematical simplifications, but on a simplification of the whole nature of the process. This simplifying assumption was that heat regeneration was governed by the same type of equation as heat recuperation. The calculation in Part III is based on this assumption.

In the recuperator equation the final preheated air temperature is expressed in terms of:

(*a*) Initial temperature of cold air
(*b*) Initial temperature of waste gas
(*c*) Heat capacity of air
(*d*) Heat capacity of waste gas
(*e*) Total heating surface
(*f*) Overall heat-transfer coefficient.

The time average of the regenerated air temperature replaces the steady final air temperature of a recuperator. Items (*a*), (*b*), (*c*), (*d*), and (*e*) occur in regenerators and recuperators, and the remaining problem is to find a value of (*f*), for regenerators. Item (*f*) is the total result of three heat-transfer processes:

(i) Heat transfer from waste gas to brick
(ii) Heat transfer from brick to air
(iii) Conduction through brick.

Various papers utilizing the recuperator equation as a basis for regenerator calculations differ only in the derived influence of (*f*)iii.

Thus Rummel[23] quotes values for the brick resistance derived mainly from empirical considerations. In the treatment suggested by Rummel, the various heat-transfer coefficients are so interrelated that a method of successive approximation should be applied to determine the effective brick resistance. However, as illustrated in Part III, the brick resistance is usually a comparatively small part of the total resistance to heat flow, and so an initial estimate is usually sufficiently accurate. Rummel includes a brief treatment of the variation of preheat level during the cycle, again largely on an empirical basis. A later paper by Southern[24] was largely a restatement of Rummel's paper, but his treatment of the variation in preheat level was based on work done after Rummel's Melchett Lecture. It is the method due to Rummel that is proposed in Part III of the report as the recommended method for calculating preheat.

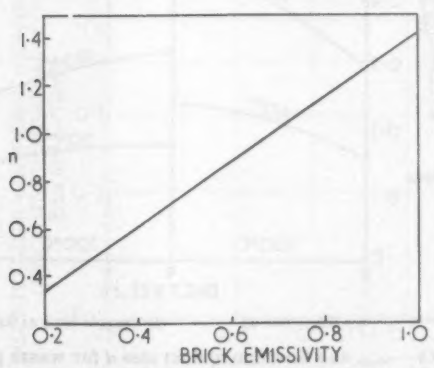


Fig. 11—Reduction factor *n*

Later publications by Hausen[25] and Schack[26] were attempts to develop methods to account for heat flow through the brick on a more purely theoretical basis.

Hausen considered solutions of the conductivity equation which would satisfy the reversal condition, i.e. after 1 gas period + 1 air period, all temperatures must return to their original values. There is an infinite number of functions which will satisfy this condition. These are known as Eigen functions. Each function is characterized by an integer known as the Eigen value, or order of the Eigen function. Hausen indicates the analogy with a vibrating string, where the zero-order Eigen function corresponds to the fundamental oscillation, and the higher-order Eigen functions correspond to the higher harmonics. Hausen defines the zero-order Eigen function as that which will yield the same temperature distribution in the direction of gas flow as occurs in a recuperator when the heat capacities of the air and waste gas are equal. Under the latter conditions, the brick temperature varies linearly in the direction of gas flow. He obtains the zero-order Eigen function satisfying these conditions and derives the corresponding overall heat-transfer coefficient. The temperature distribution in a regenerator will deviate from linear, and the deviation will be larger near the entry. The deviations are defined by the conditions that the initial air and waste-gas temperatures are constant. These conditions are satisfied by the introduction of the higher-order Eigen functions.

Hausen simplifies the general case so that deviations from conditions corresponding to the zero-order Eigen function are accounted for by applying correction factors to the heat-transfer coefficient derived from the zero-order Eigen function.

The heat-transfer coefficient for a regenerator in which the brick surface temperature varied linearly with time formed the basis of Schack's treatment.

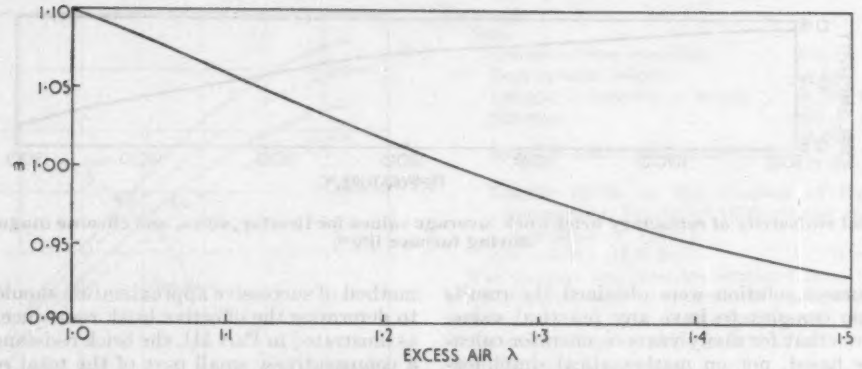


Fig. 12—Correction factor *m* for calculation of waste-gas radiation for different values of excess-air coefficient for heavy fuel oil, creosote-pitch, or tar

He defined this as the heat-transfer coefficient for the 'ideal regenerator.' He then considered the time variation of the difference between the surface temperature and mean temperature of the brick. He postulated an exponential form of this function with three arbitrary constants, two of which were determined from the boundary conditions and one of which was determined empirically. The form of the function was chosen to allow immediate integration of the derived expression for the time variation of the mean brick temperature, and no account was taken of the conductivity equation. Schack then applied similar treatment to the time variation of the difference between instantaneous gas temperature and initial gas temperature. The overall heat-transfer coefficient eventually obtained was expressed in terms of that for the ideal regenerator.

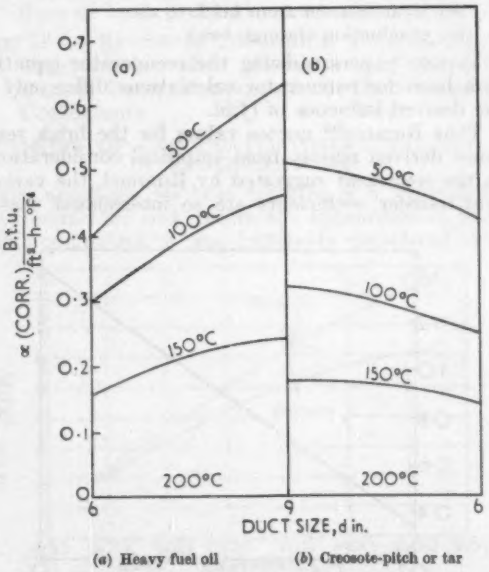
The papers mentioned above have all been intended to derive data for use in the recuperator equation. The essential difference occurs in the treatment accorded to the resistance to heat flow through the brick. Considering the present lack of accurate knowledge of the physical properties of the brick under operating conditions, and considering also that the brick resistance is a relatively small part of the total resistance (cf. Part III), the accuracy of regenerator calculations is unlikely to be greatly influenced by the selection of either Rummel's, Hausen's, or Schack's method.

Collins, Daws, and Taylor[27] developed approximations of the heat-transfer equations suitable for solution by an electronic digital computer. Their method was basically different from earlier methods inasmuch as the equations governing the process are approximated rather than the process itself. A descriptive account of their paper has been written by Burton.[28]

Collins, Daws, and Taylor assume that within the brick the temperature gradients in the direction of gas flow are comparatively small, so that conduction through the brick in that direction can be neglected. They also assume that there is no gas or air temperature gradient perpendicular to the flow direction. Using the Laplace transformation, they solve the conductivity equation, and apply the condition that no heat flows through the centre of the brick (i.e. the brick is equally heated on both sides). The initial temperature distribution is eliminated by applying the condition that after a large number of reversals all temperatures become periodic functions of time. The resulting equation relates the brick surface temperature to the temperature gradient into the brick at the surface for any point at the regenerator and for any instant.

The temperature gradient at the brick surface is related to the difference between gas (or air) temperature and brick surface temperature by a heat balance in terms of the heat-transfer coefficient waste gas to brick (or brick to air). It is also related by a heat balance to the temperature gradient of the gas (or air) in the direction of flow. By means of these relations Collins, Daws, and Taylor obtain two simultaneous equations for the gas and air temperatures. These equations contain definite integrals with respect to time and temperature gradients of the gas and air in the direction of flow.

Each of the two equations is approximated to a set of equations (which can be represented by a single equation in matrix form) by subdividing the time



(a) Heavy fuel oil (b) Creosote-pitch or tar

Fig. 13— $\alpha_{corr.}$ as a function of duct size *d* for waste gases of heavy fuel oil, creosote-pitch, or tar; parameter temperature difference between gas and brickwork

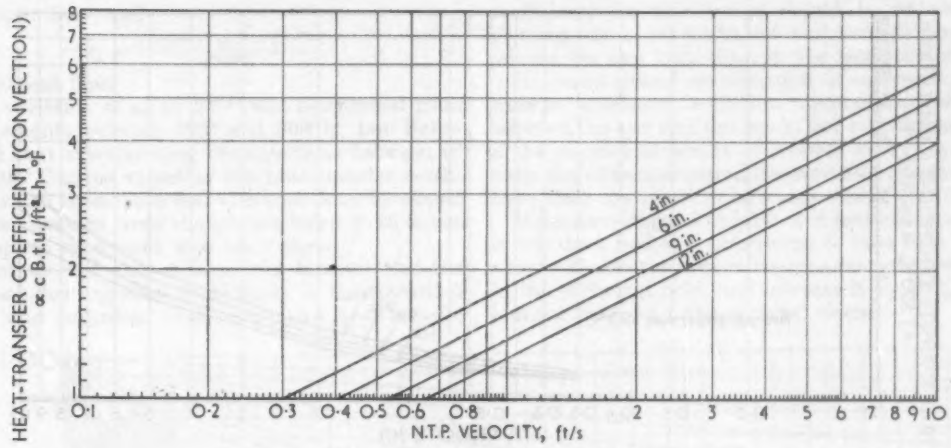


Fig. 14—Basketwork regenerator

period and applying the equations to each interval, the integration now being replaced by normal summation. The regenerator height is then subdivided, and the gas and air temperature gradients approximated to a finite difference form.

Thus at any time the gas and air temperatures at any subdivision can be represented in terms of the gas and air temperatures in an adjacent section, and so eventually can be obtained in terms of the initial gas and air temperatures, which are known. Three possible finite different forms were suggested and discussed.

The method of Collins, Daws, and Taylor is more theoretically correct than any preceding method, but suffers from the disadvantages of being less easily understood by laymen and requiring computers for its ready application. In any case our knowledge of the required data is not at present sufficiently accurate to justify this more exact treatment.

DETERMINATION OF HEAT-TRANSFER COEFFICIENTS

The accuracy of regenerator calculations depends on a knowledge of the heat-transfer coefficients as well as the correctness of the mathematical technique.

Kistner[29] described tests on an experimental basket-work regenerator to determine convective heat-transfer coefficients and pressure losses. He also developed a graphical method to simplify the calculation of radiation heat-transfer coefficients for gaseous fuels. The latter method is used in Part III, where the graphs appropriate to average modern British liquid fuels are given.

The experimental regenerator used by Kistner was 6·56 ft square by 16·4 ft high. Heat-transfer and pressure-loss tests were carried out for double staggered basketwork filling with 3·72 in. square ducts, and for straight-through basketwork filling with 3·72 in. square and 2·25 in. square ducts. Further pressure-loss measurements were made for straight-through basketwork filling with 1·2 in. square ducts.

It was immediately obvious that the double staggered filling gave the better heat transfer, as the preheat levels from this arrangement averaged 50° C higher than the preheat levels from comparable tests with straight-through filling.

In the preceding section it has been stated that the brick resistance is usually a comparatively small part

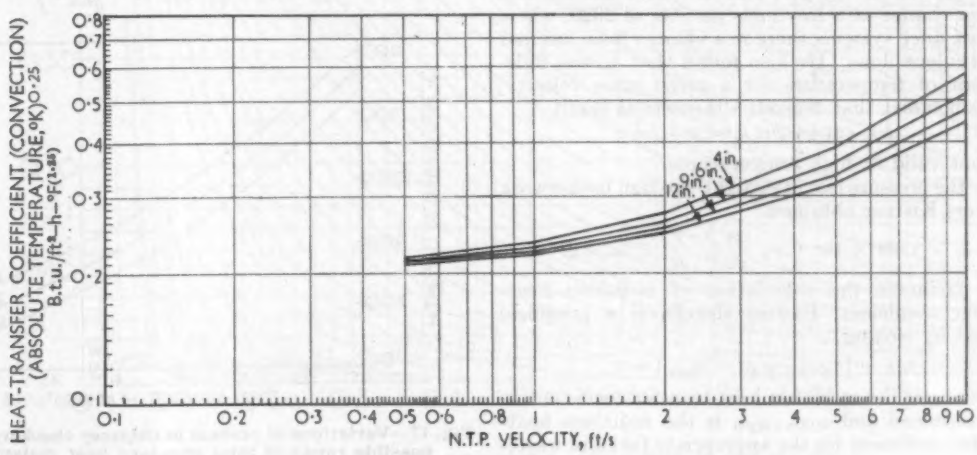
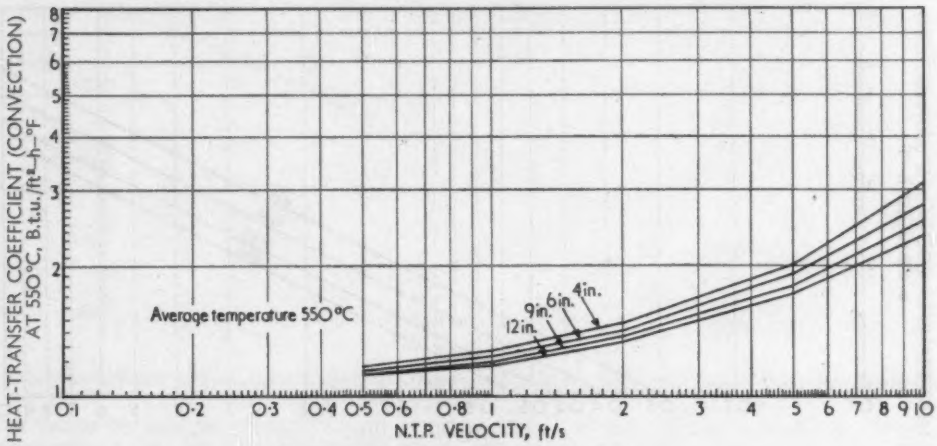


Fig. 15—Chimney regenerator



Regenerator	27 ft × 27 ft × 15 ft high	Regenerator wall thickness	3 in.
Initial waste-gas temp.	1600° C	*Curve I*	6 in. × 6 in. ducts (A = 38,500 ft²)
Initial air temp.	100° C	*Curve II*	7½ in. × 7½ in. ducts (A = 35,400 ft²)
C_gV_g	12,900 B.t.u./° F/h	*Curve III*	9 in. × 9 in. ducts (A = 32,500 ft²)
b	1·4		

Fig. 16—Chimney regenerator

of the total heat-flow resistance. As a consequence of this, it would be expected that the reversal time would have very little effect on heat transfer. This is quite consistent with the results of Kistner, who found that the total heat-transfer coefficient remained sensibly independent of the reversal time within the range of the tests (20–120 min). The reason why reversal time is so uncritical is that within the range of practical interest the heat capacity of the regenerator brickwork is many times greater than that of the air or waste gas per cycle.

Kistner eventually obtained the relation

$$a_c = \frac{2 \cdot 9 W_0^{0 \cdot 5}}{d^{0 \cdot 333}}$$

for straight-through basketwork checkers. For double staggered checkers Kistner suggested

$$a_c = \frac{3 \cdot 25 \, W_0^{0 \cdot 5}}{d^{0 \cdot 333}}$$

as a close enough approximation.

He found that the above results were independent of the nature of the gas, and also that the relation did not change at a Reynolds number of 2320, where in most duct systems there is a change from laminar to turbulent flow. He also found that a_c was independent of temperature for a given mass velocity, and concluded that Nusselt's theoretical result

$$a_c = C W^{0 \cdot 5} \gamma^{0 \cdot 5} c_p^{0 \cdot 5} \lambda^{0 \cdot 2} d^{-0 \cdot 2}$$

was not valid at high temperatures.

For the pressure loss in straight-through basketwork checkers Kistner obtained

$$h_r = 1 \cdot 39 \times 10^{-4} \cdot \gamma_0 \cdot \frac{W_0^2}{d^{0 \cdot 25}} \cdot \frac{T}{B} \cdot L \text{ in. w.g.}$$

To facilitate the calculation of radiation heat-transfer coefficient, Kistner developed a graphical method by writing

$$a_{GR} = [a_{(CO_2 + H_2O)} + a_{corr.}] \, m.n$$

where a_{GR} is the radiation heat-transfer coefficient to be determined and $a_{(CO_2+H_2O)}$ is the radiation heat-transfer coefficient for the appropriate fuel and waste-gas temperature under standard conditions for excess air, brick emissivity, and temperature difference between the waste gas and the brick; $a_{corr.}$ is a correction for deviations of the temperature difference between the gas and the brick from the standard temperature difference; m is a correction factor for deviation of the excess air from the standard; and n is a correction factor for deviation of the brick emissivity from the standard. Values of $a_{(CO_2+H_2O)}$, $a_{corr.}$, m, and n can all be conveniently obtained from diagrams, as described in Part III.

Data for convective heat-transfer coefficients for chimney-type checkers have been determined by Bohm[30] by tests on single ducts 13 ft long. The duct width was varied between 1·62 in. and 3·48 in., and the velocities between 1·3 and 23 ft/s. While technical difficulties of measurement caused some uncertainty in the field of laminar flow, Bohm considered that sufficiently accurate results were obtained for turbulent flow. He eventually obtained the working equations

$$a_c = T^{0 \cdot 25} \left(0 \cdot 198 + \frac{0 \cdot 066 W_0}{d^{0 \cdot 4}} \right)$$

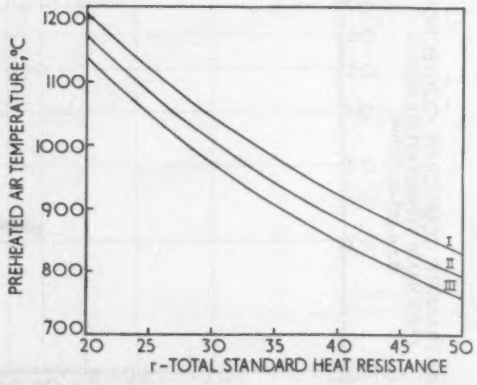


Fig. 17—Variations of preheat in chimney checkers over possible range of total standard heat resistance r for various checker fillings

for laminar flow, and

$$a_c = \frac{0{\cdot}159\ W_0^{0{\cdot}8} \cdot T^{0{\cdot}25}}{d^{0{\cdot}333}}$$

for turbulent flow.

The variation of a_c as $T^{0{\cdot}25}$ was established from measurements between 125° and 800° C, but Bohm showed that extrapolating the equations between 0° and 1200° C gives values of the heat-transfer coefficients which agree quite well with equations developed by other workers, even though the index 0·25 is not in complete agreement with their theory.

Bohm showed that neither the laminar nor the turbulent-flow equation could apply in the transition range, and suggested that convective heat-transfer coefficients for that range should be obtained by plotting values for turbulent and laminar flows and joining the two lines through the transition range.

He could detect no variation of convective heat-transfer coefficient with the temperature difference between the gas and the brick, but pointed out that, if the theoretical results of Nusselt were obeyed, the variation of temperature difference during tests would have been too small to have any observable effect.

Bohm investigated the effect of artificial roughness in the duct, and it is interesting to note that he was able to double the convective heat-transfer coefficient in the turbulent field, and increase it by 60% in the laminar field, by using serrated ducts.

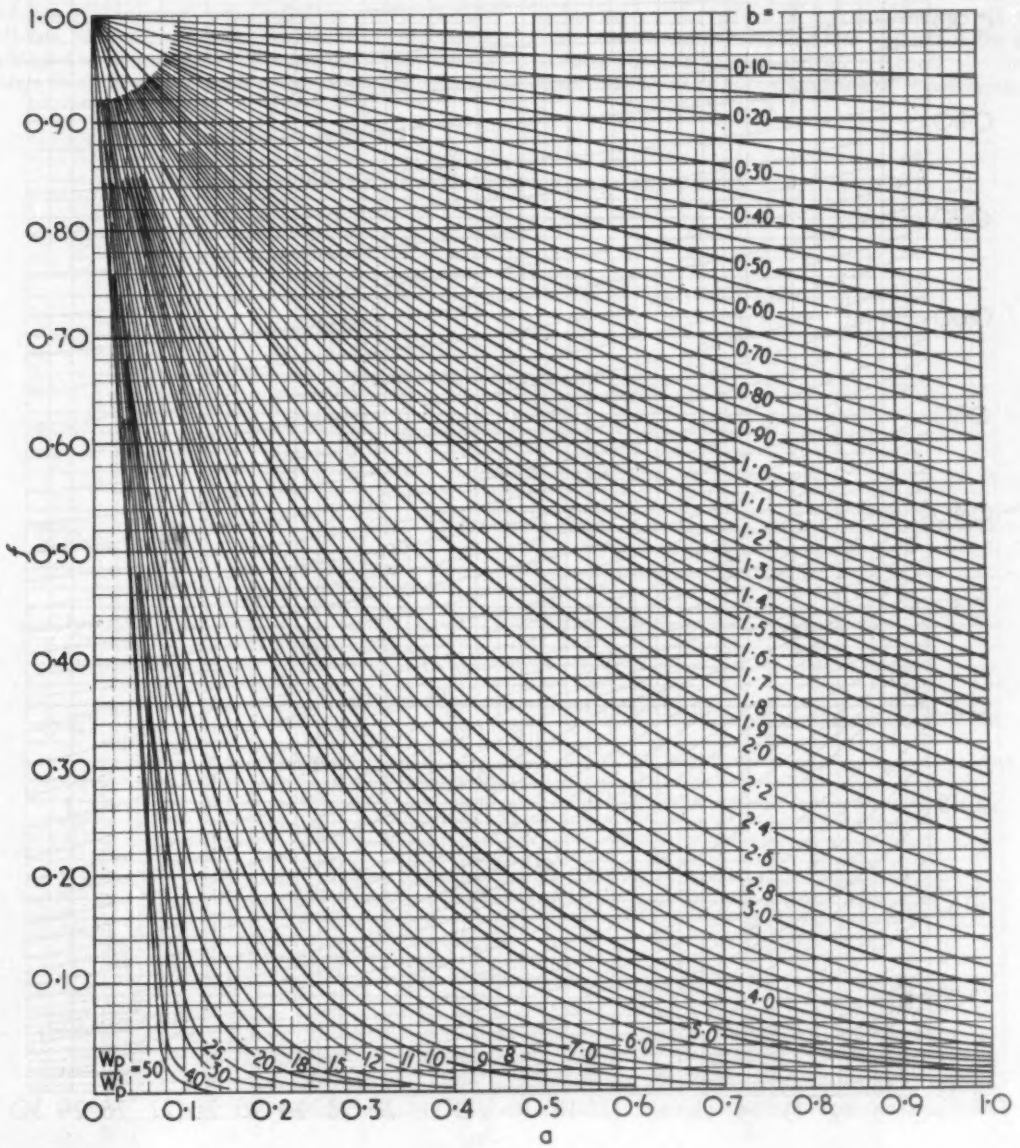


Fig. 18—Chart for function of a, b for a from 0 to 1 (Rummel[22])

VALIDITY OF PREHEAT CALCULATIONS

The most important part of the resistance to regenerative heat transfer is the transfer from brick to air. This is purely a convective mechanism, and so the effective resistance of this stage is high compared with conduction through the brick and transfer from waste gas to brick, which contains a comparatively efficient radiation heat-transfer term. The accuracy of preheat calculations will therefore depend mainly on the accuracy with which the coefficient of convective heat transfer from brick to air is known.

The equations of Kistner and Bohm are purely empirical, and so deviations from those equations may occur outside the experimental range. De Graaf and van Stein Callenfels[16] suggest that, because of the interrupted form of basketwork ducts, flow in these can probably be considered turbulent. They have therefore arranged the results of Kistner, and also the turbulent-field results of Bohm showing Nusselt numbers as a function of Reynolds numbers, in

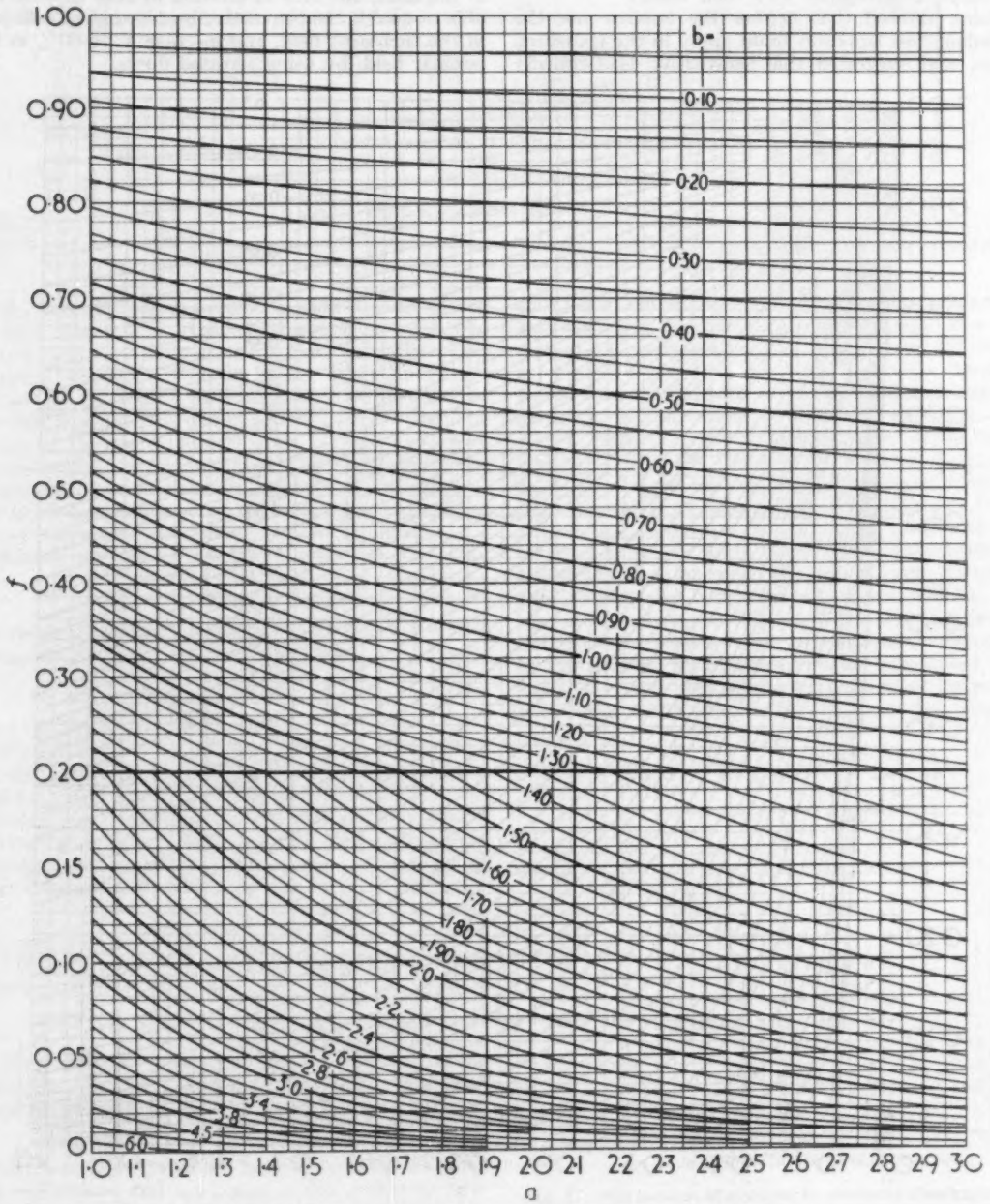


Fig. 19—Chart for function of a, b for a from 1 to 3 (Rummel[22])

accordance with similarity theory for turbulent flow. This overcomes the difficulty of the empirical nature of the original equations.

The difficulty of obtaining the correct heat-transfer coefficient for laminar flow in chimney ducts still remains. It is, however, possible to fix upper and lower limits for the heat-transfer coefficient. A lower limit can be obtained from the data of Washington and Marks,[21] since these results are for purely streamline flow. An upper limit can be obtained from the coefficient for a basketwork checker of the same duct width, since the turbulence generated by the discontinuities in the ducts of the latter will increase the heat transfer. For chimney checkers, a change of duct size, within practicable limits, in a given checker shell has little effect on the convective heat-transfer coefficient (*see* Fig. 16), but there will be a considerable change in surface area with a corresponding effect in preheat. This may be seen by examination of the results of Bohm (cf. Figs. 15 and 16) or of Washington and Marks. Within the possible range of values for the heat-transfer coefficient, the curves of preheat level against r (total standard heat resistance) for various fillings of a regenerator shell are almost parallel (*see* Fig. 17), so that the difference in preheat predicted for the different fillings will not be greatly affected by uncertainty of the heat-transfer coefficient, even though the calculated preheat levels may be seriously in error.

The flow distribution through the checker is usually assumed to be uniform for preheat calculation purposes. In most cases this condition is far from satisfied in practice. The effect of non-uniformity will vary from case to case, becoming less important for more efficient checker fillings, but as yet there are insufficient data to draw any quantitative conclusions.

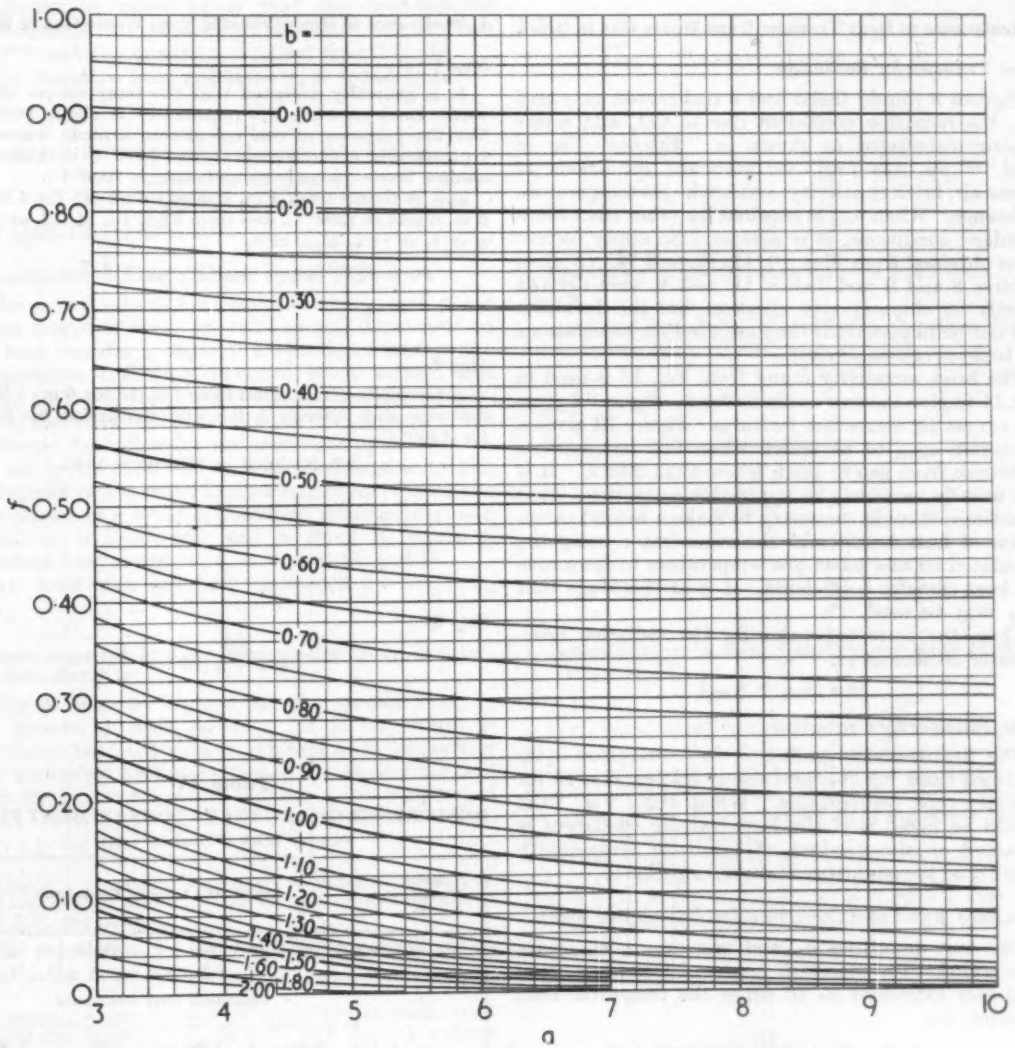


Fig. 20—Chart for function of a, b for a from 3 to 10 (Rummel[22])

Part III—A Recommended Method of Calculating the Heat Transfer in Regenerators, with a Worked Example

The following method of calculating the preheat temperature (as defined p. 255) is based on the method due to Rummel[23] given in principle in Part II (*see also* Etherington[33]). A worked example is given to demonstrate the method.

In the calculation of heat transfer in regenerators, it is necessary to find the resistance to heat flow of the regenerator and the heat capacity of waste gas and air per hour. The resistance to heat flow is found in terms of the standard heat resistance, which is the total of three terms:

(1) Resistance to heat transfer from waste gas to brick
(2) Resistance to heat transfer from brick to air
(3) Resistance to heat transfer due to the brickwork.

1. Resistance to Heat Transfer from Waste Gas to Brick

Heat Transfer by Radiation

Figures 8 (liquid fuels) and 9 (coke-oven gas) give a_{GR}, the radiation coefficient due to CO_2 and water vapour (calculated as shown in "Efficient Use of Fuel,"[22] pp. 144–150), for standard conditions of excess air, brick emissivity, and brick–gas temperature difference. When a_{GR} is required for other than these standard conditions, it is necessary to apply corrections obtained from Figs. 10, 11, 12, and 13. (*N.B.*—Figures 8 and 9 and Tables IV and V were derived strictly for chimney-type checkers, but the deviation will not be important if they are used for calculations on basketwork checkers.)

The brick emissivity found from Fig. 10 is used in Fig. 11 to give the correction factor n. Figure 12 gives the excess air correction factor m. Figure 13 gives a correction $a_{corr.}$ to be added when the temperature difference from gas to brick is less than 200° C. It is not usually necessary to apply this correction. It is sometimes thought necessary to make a second calculation of heat transfer in the regenerator using the calculated air and waste-gas temperature to determine the heat-transfer coefficients. It is at this stage that $a_{corr.}$ may be used.

Thus, the corrected value for the radiation heat-transfer coefficient is:

$$m.n\ (a_{GR} + a_{corr.})$$

Heat Transfer by Convection

The heat-transfer coefficient for convection a_c is obtained from Fig. 14 for basketwork, or Fig. 15 for chimney-type regenerators. When using Fig. 15 it should be noted that the heat-transfer coefficient is obtained as the product of (absolute temperature °K)$^{0.25}$ and the quantity obtained from Fig. 15.

Combined Effect and Correction for Interrupted Flow

The two quantities a_c, and $m.n\ (a_{GR} + a_{corr.})$ are now added. The standard heat resistance r_1 is conveniently expressed as 10 times the reciprocal heat transfer, i.e.

$$r_1 = \frac{10}{a_c + m.\ n(a_{GR} + a_{corr.})}.$$

WORKED EXAMPLE

(The worked example is given in the right-hand column, with the related description of each step in the left.)

A regenerator has chimney-type flues of 6 in. × 6 in. passages and is built of 3-in. fireclay bricks. The total exposed heating surface is 21,875 ft² (= A). Waste-gas flow is 10^6 ft³/h. Air flow is 10^6 ft³/h. Analysis of waste gas: 11·08% CO_2, 9·28% H_2O. Waste gas enters regenerators at 1600° C and leaves at 1000° C. The air enters at 100° C and a mean temperature of 550° C is assumed. Velocity of waste gas 1 ft/s (N.T.P.). Velocity of air 1 ft/s (N.T.P.). The furnace cycle is 30 min and waste gas and air periods are equal. Oil fuel used. Emissivity of brick: 0·69 at mean waste-gas temperature of 1300° C. Excess air = 25% (λ = 1·25).

1. Resistance to Heat Transfer from Waste Gas to Brick

Step 1

It is generally accepted that the temperature of the waste gases entering the regenerator is about 1600° C. For the purposes of this calculation a mean waste-gas temperature in the regenerator of 1300° C is taken (i.e. assume waste-gas exit temperature = 1000° C).

a_{GR} is found from Fig. 8 (heavy fuel oil) for 6 in. × 6 in. ducts at 1300° C, and from Figs. 10, 11, 12, and 13, n = 1, m = 1, $a_{corr.}$ = 0.

$$n.m\ (a_{GR} + a_{corr.}) = a_{GR} = 4\cdot7.$$

Step 2

a_c for waste gas is found from Fig. 15 for 6 in. × 6 in. flue size and velocity 1 ft/s and multiplied by factor for 1300° C:

$$a_c = 0\cdot23\ (1300 + 273)^{0\cdot25} = 1\cdot45.$$

Step 3

$$r_1 = \frac{10}{4\cdot7 + 1\cdot45} = 1\cdot6.$$

Table IV

RESISTANCE OF REGENERATORS TO HEAT FLOW

Fireclay Brick (Diffusivity 0·023)

Brick Thickness, in.	Length of Complete Cycle, min			
	15	20	30	40
3	Standard Heat Resistance			
	0·82	1·00	1·20	1·30

This figure for the heat resistance is not, however, the true one, as it implies continuous flow. It is necessary then to find an equivalent heat resistance which, on the basis of steady flow, would transmit the same amount of heat per hour as occurs with the actual interrupted flow. It is obvious that for continuous flow a greater resistance is required and thus the heat resistance, as determined, is multiplied by:

$$\frac{\text{Total time of cycle}}{\text{Time the waste gas flows}}$$

It is usual for the times of waste-gas and air flow to be equal and, therefore, the ratio becomes 2, i.e. the calculated heat resistance must be doubled.

2. Resistance to Heat Transfer from Brick to Air

The heat-transfer coefficient for convection a_c is obtained directly from Fig. 14 or 16. When for a chimney regenerator the average air temperature is not 500–600° C, a_c should be obtained from Fig. 15. It should be noted again that the heat-transfer coefficient is the product of (absolute temperature ° K)$^{0\cdot25}$ and the quantity obtained from Fig. 15.

The standard heat resistance r_2 is found from

$$r_2 = \frac{10}{a_c}$$

The factor for interrupted flow in this case is:

$$\frac{\text{Total time of cycle}}{\text{Time the air flows}}$$

and again this ratio is 2.

3. Resistance to Heat Transfer due to Brickwork

The actual rate of heat transfer during each phase of the complete cycle, as well as being dependent on the heat transfer coefficient, is also dependent on the temperature difference between brick surface and fluid. Owing to the imperfect heat-absorbing properties of the brick, the mean surface temperature of the brick will be higher on the waste-gas cycle than on the air cycle. The loss of heat transfer due to this divergence of the mean brick temperature during the two phases of a cycle is expressed conveniently as a resistance to heat flow and is given in terms of standard heat resistance r_3 in Tables IV and V.

Thus, total standard heat resistance $r = 2r_1 + 2r_2 + r_3$.

4. Determination of Heat Capacity per Hour of Waste Gas and Air

This is done by multiplying the specific heat by the volume of fluid flowing. It is important to remember that the volume referred to is the hourly flow multiplied by the fraction of the whole time that each fluid is flowing and is (time of flow)/(time of cycle). For equal gas and air periods this fraction is $\frac{1}{2}$; e.g. for air

$$C_a \times V_a$$

where C_a = specific heat of air at mean temperature, B.t.u./ft^3·° F (N.T.P.), and V_a = $\frac{1}{2}$ volume of air at N.T.P. per hour.

The calculation for the waste gas is carried out in exactly the same way, i.e.

$$C_g \times V_g$$

where C_g = specific heat of waste gas at mean temperature, B.t.u./ft^3·°F (N.T.P.) and V_g = $\frac{1}{2}$ volume of waste gas at N.T.P. per hour.

Table V

RESISTANCE OF REGENERATORS TO HEAT FLOW

Silica Brick (Diffusivity 0·032)

Brick Thickness, in.	Length of Complete Cycle, min			
	15	20	30	40
3	Standard Heat Resistance			
	0·80	0·90	1·1	1·2

* From Etherington,[35] pp. 343–344

2. Resistance to Heat Transfer from Brick to Air

Step 4

a_c for air is found from Fig. 16 for 6 in. × 6 in. flue size, velocity 1 ft/s, average temperature 550° C:

$$a_c = 1\cdot23$$

Step 5

$$r_2 = \frac{10}{a_c} = 8.$$

3. Resistance to Heat Transfer due to Brickwork

Step 6

r_3 is found from Table IV, for fireclay bricks at 30 min for complete cycle of waste gas and air:

$$r_3 = 1\cdot2.$$

Step 7

Total standard heat resistance:

$$r = 2r_1 + 2r_2 + r_3$$
$$r = 3\cdot2 + 16 + 1\cdot2$$
$$= 20\cdot4.$$

4. Determination of Heat Capacity per Hour of Waste Gas and Air

Step 8

It is now necessary to determine CV (the heat capacity per ° F) of waste gas and air. It is important to remember that the volume V is the hourly flow multiplied by the fraction of the whole time that each fluid is flowing, i.e. Time of flow/Time of cycle. (For equal gas and air periods this fraction is $\frac{1}{2}$).

Assume specific heat of air to be 0·020 and of waste gas to be 0·0228 B.t.u./ft^3·° F.; then $C_aV_a = 0\cdot020 \times (10^6 \times \frac{1}{2}) = 10{,}000$ (air) and $C_gV_g = 0\cdot0228 \times (10^6 \times \frac{1}{2}) = 11{,}400$ (waste gas).

5. Determination of Values of a and a' in Figs. 18–20

Having reached this stage of the calculation, the values of a and a' are needed to apply to Figs. 18–20. This is done by applying the following formula:

$$a = \frac{10 \times \text{total heating surface}}{\text{Total resistance to heat transfer} \times \text{heat capacity per degree}}$$

i.e.

$$a = \frac{10 \times A}{r.C_g V_g} \text{ for waste gas}$$

and

$$a = \frac{10 \times A}{r.C_a.V_a} \text{ for air.}$$

6. Determination of Values of b and b' in Figs. 18–20

Once again, in order to read Figs. 18–20, the values of b and b' are needed. This is done by applying the following formula:

$$b = \frac{\text{Heat capacity per hour for waste gas}}{\text{Heat capacity per hour for air}} = \frac{C_g.V_g}{C_a.V_a}$$

$$b' = \frac{\text{Heat capacity per hour for air}}{\text{Heat capacity per hour for waste gas}} = \frac{C_a.V_a}{C_g.V_g}$$

7. Determination of Factors f and f' in Figs. 18–20

The values of f and f' have been calculated using the formula:

$$f = \frac{1 - b}{1 - b \exp\left[-a\,(1 - b)\right]}$$

$$f' = \frac{1 - b'}{1 - b' \exp\left[-a'(1 - b')\right]}$$

and conveniently read off Figs. 18–20 using a with b for f; and a' with b' for f'.

8. Determination of Air Preheat and Waste-gas Outlet Temperatures

This, of course, is the final stage of the calculation and consists of applying the following formulæ:

(1) *For temperature of air leaving the regenerator*
Temp. air out = temp. of gas in − (temp. of gas in − temp. of air in) × f

(2) *For temperature of waste gas leaving regenerator*
Temp. of gas out = temp. of air in + (temp. of gas in − temp. of air in) × f'.

In the example just described, the following assumptions made were

(1) Mean waste-gas temperature
(2) Mean incoming air temperature
(3) That mean temperature difference gas–brick was 200° C.

It is now possible to determine the correct values of waste-gas temperature and incoming air temperature, and the calculation can be repeated substituting these for the values in the original example.

The new preheat and waste-gas temperature can also be used to determine the true gas–brick temperature difference by using a_e (air); $a_{GR} + a_e$ (waste gas); and total heat transfer air to brick per ft²; brick to air per ft², as follows:

5. Determination of Values a and a' in Figs. 18–20

Step 9

It is necessary to evaluate a and a':

$$a = \frac{10\,A}{rC_gV_g} = \frac{10 \times 21{,}875}{20{\cdot}4 \times 11{,}400} = \underline{0{\cdot}94} \text{ (waste gas)}$$

$$a' = \frac{10\,A}{rC_aV_a} = \frac{10 \times 21{,}875}{20{\cdot}4 \times 10{,}000} = \underline{1{\cdot}07} \text{ (air)}$$

6. Determination of Values b and b' in Figs. 18–20

Step 10

It is now necessary to evaluate b and b':

$$b = \frac{C_gV_g}{C_aV_a} = \frac{11{,}400}{10{,}000} = \underline{1{\cdot}14} \text{ (waste gas)}$$

$$b' = \frac{C_aV_a}{C_gV_g} = \frac{10{,}000}{11{,}400} = \underline{0{\cdot}88} \text{ (air)}$$

7. Determination of Factors f and f' in Figs. 18–20

Step 11

From Figs. 18–20 using (a with b) and (a' with b') we find f: and f'

$f = 0{\cdot}47$ $\left(f = \dfrac{1 - b}{1 - b \exp\left[-a(1 - b)\right]}\right)$

$f' = 0{\cdot}53$

8. Determination of Air Preheat and Waste-gas Outlet Temperatures

Step 12

(*a*) $\text{temp.}_{\text{air out}} = \text{temp.}_{\text{gas in}} - (\text{temp.}_{\text{gas in}} - \text{temp.}_{\text{air in}}) \times f.$

$\text{temp.}_{\text{air out}} = 1600 - (1600 - 100) \times 0{\cdot}465$

$= \underline{902° \text{ C}}$

(*b*) $\text{temp.}_{\text{gas out}} = \text{temp.}_{\text{air in}} + (\text{temp.}_{\text{gas in}} - \text{temp.}_{\text{air in}}) \times f'.$

$\text{temp.}_{\text{gas out}} = 100 + (1600 - 100) \times 0{\cdot}53$

$= \underline{895° \text{ C}}$

Thus preheat temperature = 902° C
Waste-gas exit temperature = 895° C.

Waste-gas to brick temperature difference

$$= \frac{(\text{Temp.}_{\text{in}} - \text{temp.}_{\text{out}}) \text{ specific heat} \times V_g}{[a_{GR} + a_c(\text{waste gas})] \times \text{total heating surface}}$$

Air to brick temperature difference

$$= \frac{(\text{Temp.}_{\text{out}} - \text{temp.}_{\text{in}}) \text{ specific heat} \times V_a}{[a_c(\text{air})] \times \text{total heating surface}}$$

The value of $a_{\text{corr.}}$ can be found from Fig. 10 and the radiation term a_{GR} now becomes

$$[a_{GR} + a_{\text{corr.}}] \times mn$$

i.e. the term $a_{\text{corr.}}$ is added to the term a_{GR} before using factors m and n.

The result now obtained will be more accurate.

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A Study of the Compatibility of some Creep-resistant Steels with Liquid Bismuth in Non-isothermal Systems

By D. W. Dawe, B.Sc., A.R.S.M., G. W. Parry, B.Sc., and G. W. Wilson, Ph.D., A.Inst.P.

SYNOPSIS

The compatibility of a series of low-carbon ferritic steels with liquid bismuth and the variation in their resistance to attack and mass transfer with composition and with particular reference to chromium content has been investigated for non-isothermal systems.

If pure bismuth alone is used, increasing the chromium content of otherwise low-alloy steels confers increasing to resistance attack by the liquid metal at temperatures up to 550° C. This can be satisfactorily explained on the basis that the protection is afforded by the presence of a thin oxide film at the steel/bismuth interface which contains increasing amounts of chromic oxide.

Of the series of steels tested, the one containing 12% Cr proved to be the most resistant to attack and mass transfer, no indication of this occurrence being observed during 2500 hours' continuous testing. It seems probable that this steel would prove a satisfactory container material for times well in excess of this.

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Introduction

LIQUID BISMUTH is not only of interest as a possible coolant for nuclear reactors, but also as the fuel carrier in the liquid-metal fuel reactor.[1] It possesses an extremely low-capture cross-section

Table I

STEEL COMPOSITIONS

Nominal Steel Composition, Makers Specification								Steel Composition as Analysed					Analyses of Plug Material from Loops			
Firth-Brown Reference	C, %	Mn, %	Mo, %	Si, %	Cr, %	V, %	Ni, %	Mn, %	Mo, %	Cr, %	Ni, %	V, %	Fe, %	Cr, %	Ni, p.p.m.	Other Elements
MOVA1	0·12 max.	0·5	0·7	0·25	...	0·25	0·2	0·48	0·75	0·26	0·25	0·44	99·99	0·01	<50	Mn, Mo } Not detected chemically
LCMO	0·12 max.	0·4–0·7	0·55	0·35 max.	0·08	...	0·2	0·52	0·59	0·05	0·12	...	99·2	0·8	<50	Mn, Mo } ,,
CRMO	0·12 max.	0·3–0·7	0·35	0·60 max.	1	...	0·2	0·48	0·62	0·81	0·11	...	99·7	0·3	<25	Mn, Mo } ,,
CRM6	0·15 max.	0·3–0·6	1·0	0·50 max.	2·25	...	0·2	0·49	0·84	2·1	0·13	...	99·77	0·13	1000	Mn, Mo } ,,
CRM5	0·12 max.	0·3–0·7	0·55	0·50 max.	5	...	0·2	0·38	0·62	4·8	0·14	...	99·77	0·84	1900	Mn, Mo } ,,

for thermal neutrons, and like other liquid metals it possesses very much better heat-transfer characteristics than gases. It has been reported[2] as a result of work carried out in the U.S.A. that low-alloy ferritic steels containing small additions of chromium can be used to contain liquid bismuth at temperatures up to 525–550° C. However, it is also known that both iron and chromium possess appreciable solubilities in liquid bismuth at these temperatures.[3] These solubilities are related to the temperature by the general equation:

$$\log_{10} (\text{concentration}) = A - \frac{B}{T}$$

where A and B are constants and T is the absolute temperature. It is therefore to be expected that these steels will pass into solution in non-isothermal systems in the region of maximum temperature and that the dissolved constituents will be deposited in regions where the temperature is lowest. The phenomena taking place in such systems have already been well described.[3, 4]

In view of the trends indicated by this earlier work it was decided to examine the influence of

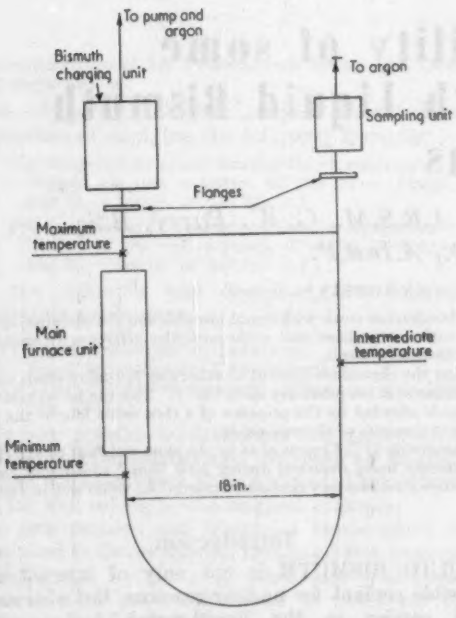

Fig. 1—Diagrammatic representation of a loop

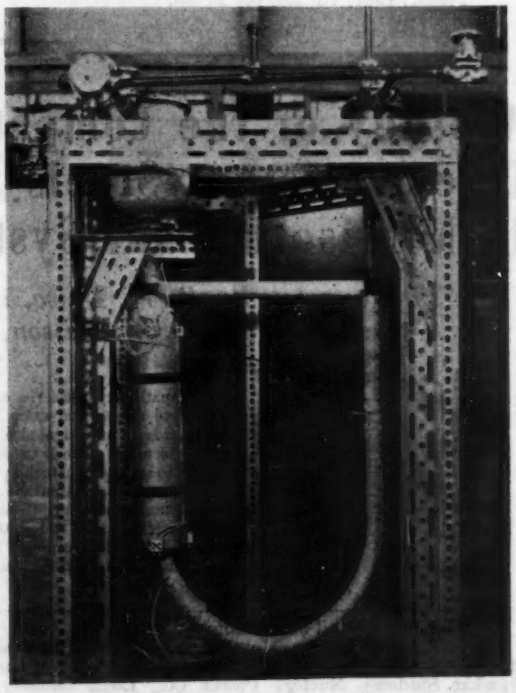

Fig. 2—Thermal convection loop in operation

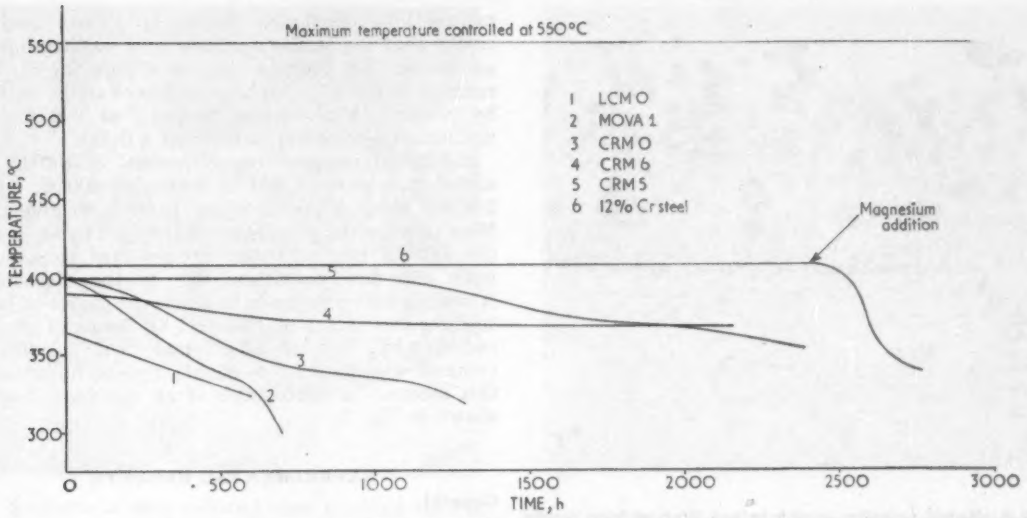

Fig. 3—Variation of temperature with time

composition, with particular reference to chromium content, on the compatibility of a series of ferritic steels at 550° C with liquid bismuth flowing in a non-isothermal system.

MATERIALS

The steels investigated were a series which had been developed for use in high-temperature steam heat-exchange systems. Since all but one of them have Firth-Brown designations, these are used throughout the paper. The nominal compositions vary from those of *MOVA*1 and *LCMO* which contain no chromium to *CRMO* containing 1%Cr, *CRM*6 containing 2–2½%Cr, *CRM*5 containing 5–6%Cr, and a steel containing 12%Cr. The nickel concentration in the steels was kept as low as possible, usually <0·2%. This stipulation was made because nickel has a very high solubility in liquid bismuth. The compositions are given in Table I.

The bismuth used was obtained from Mining and Chemical Products Ltd. and had a purity of 99·95%. The following is a typical analysis: Pb—16 p.p.m.; Sn—1·5 p.p.m.; Sb—10 p.p.m.; Cd, Zn, Co, Cu, Ni—nil; As, Ag, Au—1 p.p.m.; Fe—2 p.p.m.; Cl—19 p.p.m.

Before use the bismuth was vacuum-cast to remove oxide and particulate impurities which segregated at the top of the ingots during this process.

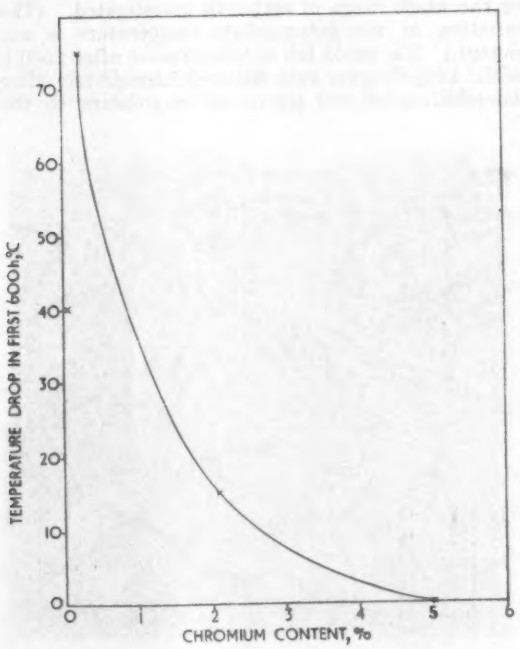

Fig. 4—Temperature drop in 600h versus chromium content

EXPERIMENTAL PROCEDURE

Tests were carried out using thermal convection loops constructed from 1 in. outside dia. by 10 or 12 s.w.g. tubing. This was bent into a U shape 18 in. in dia. and a cross-piece argon-arc welded between the uprights to form a continuous circuit. The loops were then cleaned by pickling with 20% hydrochloric acid solution to remove scale, followed by chemical polishing using a mixture of oxalic acid and hydrogen peroxide. To protect the outside of the tubing from excessive oxidation at the elevated temperatures concerned, the external surfaces were sprayed with aluminium. A loop is shown diagrammatically in Fig. 1.

The various heating circuits associated with each loop were controlled by ' Variac ' autotransformers.

The main heater consisted of six spirally wound nichrome elements sheathed in silica tubes, equally

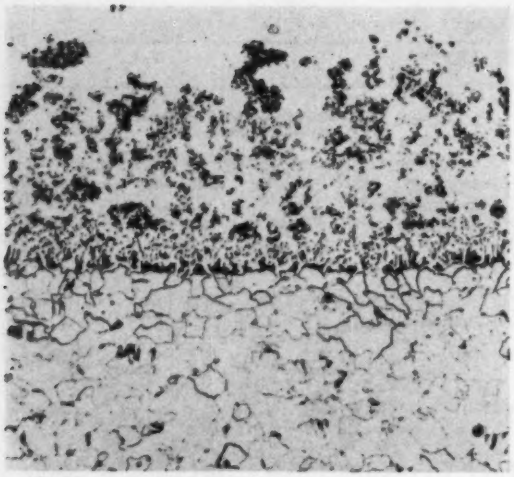

Fig. 5—Rapid solution attack in hot limb of loop made from *MOVA*1 steel. Etched in 2% Nital × 250

spaced around one limb of the loop and supported by a stainless-steel reflector lagged on its outside. The furnace was divided into two separate circuits and gave a combined heating capacity of 3 kW. One of the two circuits provided a fixed power input, the input of the other circuit being varied by means of a proportional electronic controller and pyrometer, to maintain the maximum temperature at the required value of 550° C. This temperature was measured by Pt/Pt–10%Rh thermocouple silver-soldered to the wall of the loop just above the top of the main heating unit. To ensure that during the filling operation the bismuth did not freeze and to provide, if necessary, trace-heating during the test, the rest of the loop was wound with a 1½-kW heater element of 'Brightray C' wire. A total of three Pt/Pt–10%Rh thermocouples were silver-soldered to the wall of the loop. One recorded the maximum (control) temperature, another the minimum, and, provided that bismuth was flowing round the loop, the third thermocouple recorded a temperature intermediate between the maximum and minimum. All temperatures were recorded on multipoint temperature recorders.

The loops were filled with liquid bismuth from charging units which consisted of evacuable stainless-steel pots heated by windings of 20 s.w.g. nichrome. Vacuum-tight seals were obtained between these units and the loops by using annealed aluminium O-rings between flanges on the pots and flanges brazed to the top of the loops. A sampling device was similarly attached to each loop but was not used very much for reasons explained later.

When sufficient bismuth had been placed in the charging unit the whole loop assembly was evacuated to less than 5μ Hg in order to test for leaks, and the temperature of the tubing raised to approximately 400° C. The loop was then flushed with hydrogen for 6–8 h in order to reduce any iron oxide present, the apparatus re-evacuated in order to remove hydrogen from the steel, and the bismuth in the charging unit melted into the loop. Radiography was used to ensure that the correct bismuth level was attained in all loops. To prevent ingress of air during the running of the loop, an atmosphere of argon, purified by passing over calcium turnings at 600° C, was maintained above the bismuth at 5 lb/in².

An initial temperature differential of 150° C was aimed at, maximum 550° C, minimum 400° C, trace-heating being adjusted where possible to give this. Mass transfer and plugging were detected by observing the fall in the minimum temperature consequent upon the flow restriction due to the deposition. Attempts were also made to assess the degree of mass transfer and attack on the steel by means of gamma radiography, but it was found that insufficient contrast was obtained to enable it to be followed by this means. A photograph of an operating loop is shown in Fig. 2.

EXPERIMENTAL RESULTS

General

Six loops, one each of *MOVA*1, *LCMO*, *CRMO*, *CRM*6, *CRM*5, and 12% Cr steel were operated. Initial flow velocities of the bismuth in the loops of about 1·4 cm/s were calculated from relationships connecting power input and the temperature rise.

In general, the minimum temperatures were allowed to fall to approximately 300° C by which time it was estimated that the flow velocity had fallen to 0·8 cm/s. The heating was then switched off and the bismuth was allowed to cool *in situ*. Sections were taken from each loop for metallographic examination and also for chemical analysis. The variations of the minimum temperature with time are shown in Fig. 3 for the whole range of materials investigated. (The variation in the intermediate temperature is not plotted.) The rapid fall in temperature after 2500 h in the 12% Cr steel loop followed immediately after the addition of 500 p.p.m. of magnesium to the

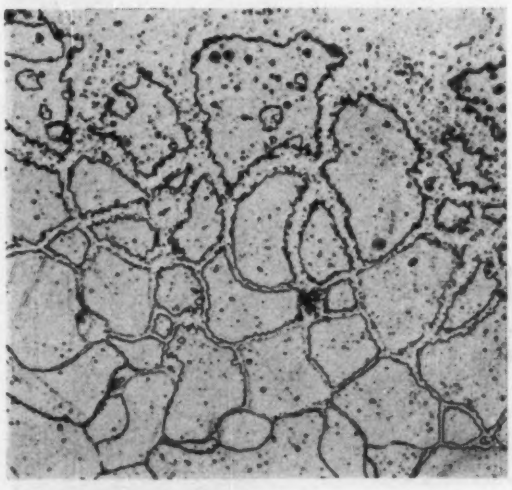

Fig. 6—Intergranular attack on *CRM*5 steel in section taken from hot limb. Etched in 2% Nital × 500

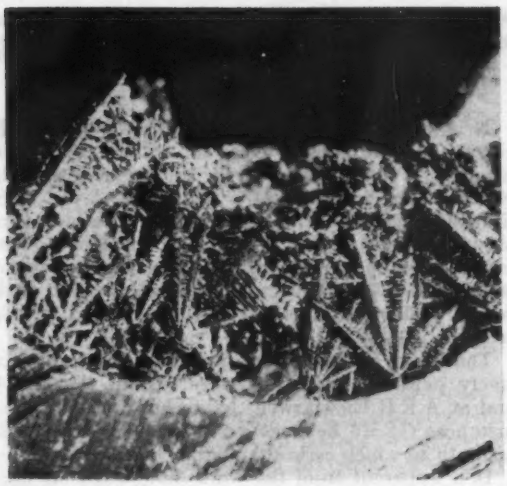

Fig. 7—Form of plug revealed after leaching out bismuth with nitric acid × 8

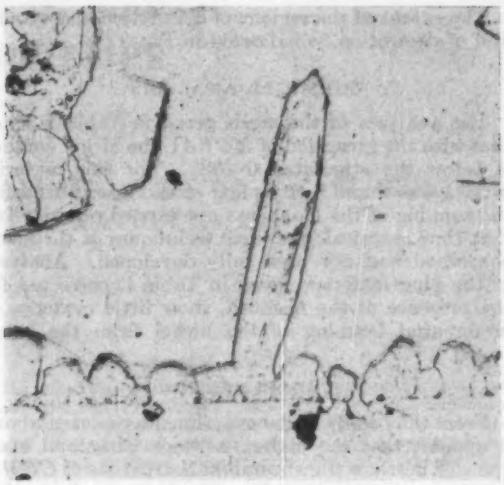

Fig. 9—Growth of dendrite from tube wall in plugged region. Etched in 2% Nital. Direction of bismuth flow from left to right

bismuth. The graphs show that the presence of chromium in the steels apparently served to decrease the rate of dissolution and mass transfer. This is better shown in Fig. 4, where the chromium content of the steels is plotted against the temperature fall occurring in the first 600 h. As can be seen from this graph, the temperature drop is a maximum for the loop constructed from *MOVA*1 and a minimum for *CRM*5 and the 12% Cr steel. The departures from the curve are probably due to initial differences in the state of the internal steel surfaces.

METALLOGRAPHY

The metallographic examination revealed that dissolution of the container wall, reaching a maximum in the region of maximum temperature, proceeded in all cases by general solution of the steel surface and by intergranular penetration. It was particularly noticeable that, with the steels containing 1–6% Cr, intergranular penetration was much more marked than with *MOVA*1, *LCMO*, and 12% Cr steels. In these latter three materials the general solution type of attack was apparently keeping pace with the bismuth penetration. Variation in the type of attack is well shown in Figs. 5 and 6. Despite the presence of many very small iron crystallites close to the tube wall in the hot limb, these must dissolve very rapidly in the main bismuth flow because it was impossible to find any particulate iron in the bismuth in specimens taken from regions other than where deposition had taken place.

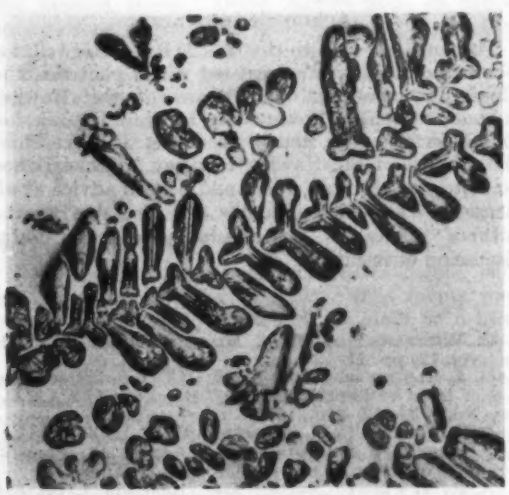

Fig. 8—Dendrites from plugged region of loop. Etched in 2% Nital × 50

Deposition of the material transferred by the bismuth occurred over a length of approximately 18 in. in all cases and was a maximum just before the position where the minimum temperature was measured. It was evident that in some cases deposition began with the formation of a thin layer of crystallites, firmly adherent to the tube wall. In other places this layer was not present at all. From some of the deposited crystallites, or from parts of the tube wall, crystallites grew out into the main bismuth stream assuming a typical dendritic form, as shown in Figs. 7 and 8. Presumably the preferential growth of crystals keyed to the tube wall, as shown in Fig. 9, originates in an orientation effect, as it is known that, in the same environment, crystal growth is very much more rapid in some crystal directions than in others. When growth has proceeded to a certain extent, the direction of growth will be governed by the higher concentration of dissolved material in the centre of the bismuth stream away from the depleted, relatively stagnant, liquid metal close to the tube wall.

The extent of the regions of dissolution and attack, and of deposition, is indicated in Fig. 1.

CHEMICAL ANALYSIS

The analyses of the steels given in Table I, show that with the exception of *MOVA*1, the nickel content is below the stipulated 0·20%. For this material 0·25% was found. Sampling of the bismuth during the running of the loops was not carried out since at that time chemical analytical techniques at the levels concerned had not been fully developed. Analyses of the plug materials given in Table I, corrected for the presence of the bismuth, show little evidence of preferential leaching of the nickel from the steels tested.

DISCUSSION

From the results of the experiments reported above it appears that the materials which withstand mass transfer best are the chromium-bearing steels *CRM*6, *CRM*5, and 12% Cr. Despite this, metallographic examination indicates that the chromium-containing steels *CRMO*, *CRM*6, and *CRM*5 are very much more susceptible to intergranular attack than those with higher or lower chromium contents. It is hoped that further comparisons will be possible as a result of stress corrosion and other work at present in hand. The minimum temperature of the loop made from 12% Cr remained steady for 2500 h, which suggests that very little transfer occurred in this period, but after the addition of 500 p.p.m. magnesium to the bismuth* very rapid mass transfer and plugging followed almost immediately (*see* Fig. 3).

The resistance of the chromium-containing steels to bismuth attack is at first sight puzzling, because it is known that pure chromium is at least as soluble as iron in bismuth. However, there appear to be at least two possible explanations of this improved behaviour:

(i) The chromium present tends to form a compound with the iron, thus increasing the energy required before atoms of the latter can pass into solution

(ii) A protective film containing chromic oxide is present at the interface between the steel and the bismuth.

The phase diagram of the Fe–Cr system suggests that the iron–chromium compound is not particularly stable. Furthermore, the behaviour of the 12% Cr steel loop indicates that the second mechanism is more probably correct, since magnesium, even in dilute solution in the bismuth, would on thermodynamic grounds reduce any chromic oxide film present on the steel and allow solution attack to proceed. Decreasing the quantities of chromium present in the alloys would be expected to lead to increased susceptibility to bismuth attack, as the oxide films would contain less of the relatively stable and protective chromic oxide.

Such films might also be expected to exhibit weak points, corresponding to underlying grain boundaries. Anisotropic thermal expansion of the steel grains would rupture such a film at these grain boundaries, leading to intergranular penetration.

* Magnesium has a reported solubility in bismuth of 3 wt-% at 440° C.

CONCLUSIONS

It is postulated that the superior resistance of the chromium-containing steels to attack by liquid bismuth is primarily due to the presence at the bismuth/steel interface of a thin oxide layer. The protective nature of this film increases with increasing chromic oxide content. However, tendency toward iron–chromium compound formation may well retard solution attack, particularly of high-chromium steels whose protective films have been ruptured or destroyed.

The above remarks only apply in the case of relatively pure bismuth. Investigations in the U.S.A. and at A.E.R.E., Harwell, have shown that in some instances *CRM*6 behaves very well in contact with bismuth to which certain additions have been made.[5]

It is apparent from the work described that none of the selected steels can be considered with certainty as a suitable container material for liquid bismuth in non-isothermal systems. Increasing the chromium content of otherwise low-alloy steels confers increasing resistance to attack by the liquid metal at temperatures up to 550° C. This can be satisfactorily explained on the basis that the protection is afforded by the presence of a thin oxide film at the steel/bismuth interface which contains increasing amounts of chromic oxide. Metallographic examination suggests that this film is sufficiently weak in the case of *CRMO*, *CRM*6, and *CRM*5 to be broken at grain boundaries, leading to severe intergranular attack.

In the case of the steel containing 12% Cr there is no evidence that film rupture occurred before the addition of magnesium to the bismuth. This steel might prove to be a satisfactory container material for bismuth, provided that sufficient oxygen is available in the bismuth to re-form the film if breaks occur.

Acknowledgments

The authors wish to thank A.E.R.E., Harwell, for whom the work was carried out under contract, the Chief Executive and the Board of the Hawker Siddeley Nuclear Power Co., Ltd. for permission to publish the paper, and Mr. G. J. Metcalfe, the Chief Metallurgist of the Hawker Siddeley Nuclear Power Co., Ltd. for encouragement and helpful advice at all stages of the work. Thanks are also due to the authors' colleagues for assistance and for useful discussion of results.

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Variation in Fatigue Properties over Individual Casts of Steel

PART II — INVESTIGATION OF A CAST OF STEEL TO B.S. 970 SPECIFICATION En 24

By E. Ineson, B.Sc., F.I.M., J. Clayton-Cave, and R. J. Taylor, B.Sc., F.S.S.

SYNOPSIS

The variations in fatigue properties (*a*) within an ingot, and (*b*) between ingots of the same cast of steel, have been determined for a typical cast of basic electric steel to B.S. 970 specification En 24. Two conditions of heat-treatment were investigated, i.e. hardened and tempered respectively to nominal tensile strengths of 85 tons/in² and 125 tons/in². For each condition 500 test-pieces were selected to represent various portions of an ingot and different ingots in the cast. These were tested in fatigue according to a statistically designed experiment. Supplementary metallurgical information concerning composition variations, distribution of non-metallic inclusions, hardenability, and tensile properties was also accumulated. It has been shown that at both levels of tensile strength the longitudinal fatigue properties were fairly uniform throughout this cast of steel. The non-metallic inclusions were of the normal types and sizes usually observed in this class of steel, although this particular cast was perhaps cleaner than average. No direct association could be detected between the incidence of fatigue failure and the presence of non-metallic inclusions, but there was some indication that certain inclusions may have contributed to the initiation of a few failures of material in the 125-tons/in² condition. The fatigue properties of the material with a tensile strength of 125 tons/in² were inferior to those of the same material with a tensile strength of 85 tons/in². 1579

INTRODUCTION

EARLIER WORK[1] on an acid open-hearth cast of steel to B.S. 970 specification En 100, heat-treated to a nominal tensile strength of 60–65 tons/in², concluded that the longitudinal fatigue properties were fairly uniform, both within an ingot and between ingots of the same cast. The fatigue limit of material from the top of an ingot was found to be about 5% higher than that of material from the middle and bottom of the same ingot, but this difference, though statistically significant, was considered to have no practical importance. Further, it was shown that any small observed variations in fatigue properties were closely linked with variations in the tensile strength, the hardness, and the chemical segregation of the material; no association was observed between the incidence of fatigue failure and the presence of non-metallic inclusions.

It was considered, however, that metallurgical variables originating from the initial cast structure of an ingot might have a more important effect on fatigue properties of steel heat-treated to higher tensile strengths. There is some evidence to show, for instance, that at higher tensile strengths, non-metallic inclusions have a more marked effect,[2–5] dependent, in part, on their orientations with respect to the applied stress. Other factors such as micro-segregation might also play a more important part at the higher strengths.

The object of the present investigation was, therefore, to study the variations in fatigue properties of a commercial cast of steel to B.S. 970 specification En 24, rolled and heat-treated to nominal tensile strengths of 85 and 125 tons/in² respectively.

METALLURGICAL DETAILS OF MATERIAL

Supply of Material

The material used came from three ingots, each weighing approximately 25 cwt. The ingots were chosen from a typical cast of steel made in a basic electric-arc furnace to B.S. 970 specification En 24. The whole cast consisted of approximately 250 cwt of steel which was teemed directly into ten moulds through a $\frac{7}{8}$-in. nozzle. Details of the ingot are shown in Fig. 1 and the selection of ingots 1, 5, and 10 for this investigation is shown in Fig. 2 with details of the stages of hot working down to $\frac{7}{8}$-in. dia. bars.

After cooling, the ingots were stripped and charged directly into an annealing furnace in which they were 'step annealed.' Following this heat-treatment, the surfaces of the ingots were removed by oxy-acetylene scarfing. The ingots were then charged into a continuous reheating furnace, heated to rolling temperature, and rolled to 5-in. square billets. At this stage the rolling reduction was 90·5% for the top of an ingot and 84·8% for the bottom. Top and bottom discards of approximately 15% and 2% respectively were then taken from the products of each ingot and the remainder was divided into five equal portions of 5-in. square billets for each ingot.

The first, third, and fifth portions from the 1st ingot and the third portions from the 5th and 10th ingots were then reheated to rolling temperature, rolled to 2-in. square billets, and then to $\frac{7}{8}$-in. dia. bars. This involved a further rolling reduction from the 5-in. square stage of 98·5%.

At all stages of rolling between ingot and $\frac{7}{8}$-in. bar, examination was made for defects. Great care was taken to ensure that the identity of bars with respect to the original ingot and ingot position was known.

Paper MG/M/215/57 of the Fatigue and Inclusions Committee of the Metallurgy (General) Divisional Panel of B.I.S.R.A., received 8th May, 1958. The views expressed are the authors', and are not necessarily endorsed by the Committee as a body. Part I of this investigation was published in the October 1956 issue of the *Journal* (vol. 184, pp. 178–185).

Mr. Ineson is Laboratory Head and Mr. Clayton-Cave an investigator in the Metallurgy (General) Division of B.I.S.R.A. Mr. Taylor is in the Operational Investigations Section of the Association.

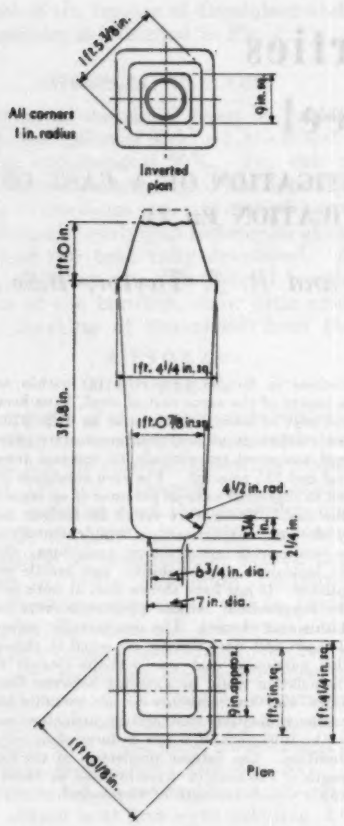


Fig. 1—Details of ingot

Table I

CHECK ANALYSIS ON SAMPLES FROM THE FIVE INGOT POSITIONS, wt-%

Element	Specification B.S. 970 En 24	Cast Analysis	Ingot 1			Ingot 5	Ingot 10
			Top	Middle	Bottom	Middle	Middle
C	0·35–0·45	0·40	0·42	0·42	0·40	0·41	0·41
Si	0·10–0·35	0·30	0·29	0·29	0·28	0·29	0·30
Mn	0·45–0·70	0·56	0·59	0·60	0·59	0·59	0·59
S	0·05 max.	0·012	0·012	0·012	0·012	0·012	0·012
P	0·05 max.	0·015	0·013	0·014	0·014	0·012	0·014
Cr	0·90–1·40	1·06	1·05	1·05	1·04	1·07	1·06
Ni	1·30–1·80	1·46	1·47	1·48	1·47	1·47	1·46
Mo	0·20–0·35	0·21	0·20	0·20	0·20	0·21	0·20

Examination of Composition, Structure, and Mechanical Properties

Analysis and Segregation—The cast analysis together with check analyses (average of five results) for the five ingot positions investigated are given in Table I. These results show that the steel is well within specification and that there is no appreciable variation in composition between the ingot positions.

The samples reported on in Table I represented the full cross-section of the bars. Previous work[1] showed that even when these sample analyses indicate complete composition uniformity, appreciable variation can still exist both from centre to outside of the section and from bottom to top of the ingot centre. This possibility was therefore investigated for the present steel by taking samples with a 0·3-in. dia. drill from the centre of $\frac{7}{8}$-in. dia. bars representing the five ingot positions. Corresponding samples were also turned from the outsides of the bars. Analyses of all these samples gave results agreeing closely with those shown in Table I, indicating that there were not significant variations in composition throughout an ingot or from one ingot to another. Sulphur prints taken from sections of 5-in. billet representing material immediately adjacent to that selected for the fatigue tests showed no segregation whatsoever, thus confirming the analytical results.

Microscopic examination of longitudinal sections showed a certain amount of banding but this was considered to be not more than is normal for this class of steel.

Response to Hardening—Conventional Jominy end-quench tests were undertaken on samples representing the five ingot positions. A sub-standard test-piece having a diameter of $\frac{3}{4}$ in. was used because the only material available was in the form of $\frac{7}{8}$-in. dia. bars. The results showed that no marked variation existed for the five ingot positions and that the hardenability curves were similar to those reported elsewhere[6, 7] for similar steels.

Thin discs cut from the bars and isothermally transformed showed no variation in the rate of transformation between the centre and the outside of the bars.

Grain Size—No difference existed between the austenitic grain size of samples taken from the five

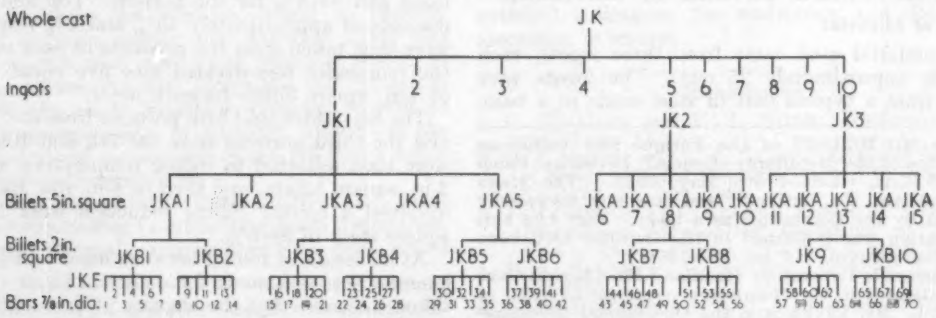

Whole cast
JK
Ingots
1 2 3 4 5 6 7 8 9 10
JK1 JK2 JK3
Billets 5in. square
JKA1 JKA2 JKA3 JKA4 JKA5 JKA6 JKA7 JKA8 JKA9 JKA10 JKA11 JKA12 JKA13 JKA14 JKA15
Billets 2in. square
JKB1 JKB2 JKB3 JKB4 JKB5 JKB6 JKB7 JKB8 JK9 JKB10
JKF
Bars 7/8in. dia.

Fig. 2—Details of cast

Table II

INCLUSION COUNTS

Position in Ingot	Type of Inclusion	Total for Range of Sizes of Widths, mm (× 0·0035)					Maximum Length, mm (× 0·0035)	General Remarks
		0–1·0	1·0–2·0	2·0–4·0	4·0–8·0	8·0–16·0		
Ingot 1 Top	Aluminium oxide	16					1·0	***Manganese sulphides*** Elongated, sometimes associated with silicate inclusions. General survey indicates MnS inclusions to be most numerous at top of ingot, but thickest inclusions found in bottom and middle portions
	Silicates	405	78	13			7·0	
	Manganese sulphide	252					35·0	
Middle	Aluminium oxide	9					1·0	***Siliceous type*** Two main varieties present: deformable and undeformable; the latter present either as spheres or as angular inclusions. Often the angular inclusions were associated with the deformable variety. Some of the spherical undeformable inclusions had 'tails'. Sometimes MnS was found with both these varieties of siliceous inclusion.
	Silicates	289	52	6			7·0	
	Manganese sulphide	146	4				20·0	
Bottom	Aluminium oxide	13					1·0	
	Silicates	212	57	10	3		8·0	
	Manganese sulphide	142	3				20·0	
Ingot 5 Middle	Aluminium oxide	...						***Aluminium oxides*** Angular particles found as clusters or as stringers with a tendency to form 'comet-tails'
	Silicates	196	41	1			10·0	
	Manganese sulphide	209					30·0	
Ingot 10 Middle	Aluminium oxide	...						
	Silicates	201	31	12	...	1	6·0	
	Manganese sulphide	200					40·0	

Each total count for a given ingot position is drawn from the examination of 120 fields at a magnification of × 360 and field of view of 0·35 mm diameter

ingot positions studied; the actual grain size of the material varied from 7 to 8 A.S.T.M. as determined from prepared micropolished samples given a brief oxidizing treatment at the austenitizing temperature.

Non-metallic Inclusions—Studies of the inclusions were made on samples taken from the critical portions of the fractured test-pieces, using a method of inclusion-counting similar to that employed in the earlier investigation.

Three samples were taken for each ingot position under investigation. The samples, approximately ½ in. long, were sectioned longitudinally and carefully polished with diamond paste to ensure that all inclusions were retained. Forty random fields were then selected for each sample, i.e. 120 fields for each ingot position, making 600 fields in all. All fields examined were confined to a region within 2·0 mm of the edge of the section since it was felt that only inclusions in this region could have any bearing on the fatigue tests. The counts are given in Table II.

It will be seen that material from the top of ingot 1 has a rather higher inclusion content, particularly of silicates, than the remaining four positions, but the majority of inclusions were very small. The larger inclusions were predominantly silicates. Aluminium oxide inclusions appear to be confined to ingot 1.

The manganese sulphides were all elongated in the direction of rolling. The silicate inclusions occurred

Table III

TENSILE TESTS REPRESENTING THE FIVE INGOT POSITIONS
Average of Six Tests

Position in Ingot	85 tons/in² Condition						125 tons/in² Condition					
	U.T.S., tons/in²	Proof Stress, tons/in²			Elongation % on $4\sqrt{A}$	Reduction in Area, %	U.T.S., tons/in²	Proof Stress, tons/in²			Elongation % on $4\sqrt{A}$	Reduction in Area, %
		0·1%	0·2%	0·5%				0·1%	0·2%	0·5%		
Ingot 1 Top	86·1	77	78	81	16	47	123·5	90	97	107	14	43
Middle	85·9	77	78	81	16	47	123·7	91	97	108	14	42
Bottom	85·2	77	78	81	16	48	121·8	90	97	108	14	45
Ingot 5 Middle	86·1	78	79	81	16	45	125·0	93	100	109	14	41
Ingot 10 Middle	86·5	77	78	81	16	46	123·9	93	100	109	14	42

Table IV

ANALYSIS OF VARIANCE
Tensile Tests (U.T.S.)

Source of Variation	Degrees of Freedom	85 tons/in² Condition		125 tons/in² Condition	
		Mean Square	Test Ratio	Mean Square	Test Ratio
Between ingots 1, 5, and 10	2	1·23	5·9*	9·18	18·0†
Between top, middle, and bottom of ingot 1	2	1·26	6·0*	6·41	12·6†
Residual	25	0·21	...	0·51	...
Total	29	...	...	...	...

* Significant, 1% level † Highly significant, 0·1% level

in two main varieties: those which were quite angular and those which were either spherical or slightly elongated with 'tails' similar to those encountered by Frith.[2] These two varieties were found in all size ranges from 0·0035 mm to 0·056 mm. All the aluminium oxide inclusions observed were angular in appearance.

Analysis of the inclusion counts resulted in evidence against the theory advanced by Palazzi[8] that non-metallic inclusions are distributed randomly within samples of steel, when these are small relative to the ingot size. In fact, generally, the presence of the inclusions was found to be associated with micro-segregation as disclosed by banding.

Tensile Properties—Tensile tests were undertaken on pieces representative of the five ingot positions, in two different conditions of heat-treatment corresponding respectively to nominal tensile strengths of 85 and 125 tons/in². The preparation of pieces and conditions of test were as follows:

(i) *Nominal 85 tons/in²*—Six pieces of ¾-in. dia. bar for each ingot position were heated in one batch to 850° C, held for one hour, and oil-quenched. They were then tempered at 500° C for one hour. Standard tensile test-pieces of 0·564 in. dia. with a 2·0-in. gauge length were machined from the blanks and tested in conformity with B.S. 18 using a 35-ton Amsler machine and a Lindley extensometer.

(ii) *Nominal 125 tons/in²*—Six pieces of ¾-in. dia. bar for each ingot position were rough machined to within 0·030 in. of final dimensions. The rough machined pieces were heated in one batch to 850° C, held for one hour and oil-quenched, followed by tempering at 200° C for eight hours. Machining to final dimensions was done by cylindrical grinding. This ensured that the results would relate to the fatigue test-pieces which were prepared similarly. A sub-standard diameter of 0·357 in. was adopted, i.e. 0·1 in² cross-sectional area, to keep the tests within the capacity of the testing machine. A Lindley or a Huggenberger extensometer with a gauge length of 2·0 in. was used for determinations of proof stress, but percentage elongations were calculated from measurements made on a gauge length equal to $4\sqrt{A}$, where A was equal to the original cross-sectional area.

The tensile test results are given in Table III. There are no major differences between the results from material representing each ingot position for either heat-treatment condition, but generally the tensile strength of the material from the bottom of ingot 1 is lower than that from the middle and top of this ingot, and the tensile strength of material from the middle of ingot 1 is lower than that from the middle of ingots 5 and 10. The statistical significance of these inferences is demonstrated by analysis of variance (U.T.S.) given in Table IV.

DESIGN OF FATIGUE EXPERIMENTS

Two experiments are discussed in the present paper and the plan for each is based on that used in an earlier investigation. For each experiment, and for each ingot position, 100 tests were intended, 10 at each of 10 stress levels. Their arrangement specified that, from each of the 10 bars in an ingot position, one and only one test-piece is used at any given stress level by each of 10 testing machines. Because of practical difficulties, described later, the initial plan had to be modified slightly; nevertheless the original aim was retained: to make the final comparison between ingot positions independent of testing machines, bar to bar variations (within an ingot position), variations along a bar, and variations in batch heat-treatment.

The 1000 test-pieces (excluding spares) were numbered consecutively from the beginning to the end of the cast. The investigation into material, 85 tons/in² condition, used only odd-numbered pieces and the investigation into material, 125 tons/in² condition, used the even-numbered pieces.

Choice of Stress Levels and Threshold Values

The choice of the ten stress levels used in the two experiments was based, in each case, on conventional S/N diagrams determined from the appropriate heat-treated material representing each of the five ingot positions.

Fatigue-limit estimates must be sensitive to changes in the threshold criterion used to decide whether a test result can be classified as a non-failure, i.e. the maximum number of cycles to be completed in any test, even when the specimen does not fracture. The choice of the threshold value, 40×10^6 cycles, for testing the 85 tons/in² material, was based on past experience. For higher tensile steels Frith[2] states that testing should be continued to at least 100×10^6 cycles; in fact the results from the 125 tons/in² material indicate that even with a threshold value of 1000×10^6 cycles an appreciable chance would still exist that the classification of a test would have to be changed from non-failure to failure if testing were continued. Nevertheless the final choice of threshold value was 100×10^6 cycles as limitations in time and in testing facilities prohibited the use of a greater number of cycles.

Heat-treatment of Test Bars

The pieces required in the 85 tons/in² condition were heat-treated in the ¾-in. dia. bar form by oil-quenching from 850° C and tempering for 1 h at 500° C. The pieces required in the 125 tons/in² condition were first rough machined to within 0·03 in. of final size and then oil-quenched from 850° C, followed by tempering for 8 h at 200° C.

For both conditions the 300 test-pieces representing the top, middle, and bottom positions of ingot 1 were

made up into ten batches of 30 pieces each, each batch containing all the pieces destined for testing at a particular stress level. The 200 pieces representing the middle portions of ingots 5 and 10 were divided into ten batches of 20 pieces each, again with each batch containing all the pieces to be tested at a particular stress level. Throughout the heat-treatment great care was taken to ensure that the conditions were similar for each batch of test-pieces.

As in the previous work, a check was made on the uniformity of the batch heat-treatments; every heat-treated test-piece was checked for hardness both before and after machining. One batch of 85 tons/in² material was found to be off-standard; after machining the average diamond hardness (HD30) was 446 D.P.N. against 418 D.P.N. for the other batches, but before machining the Brinell test results were normal. Subsequent examination showed that an increase in surface hardness had occurred during machining; as will be seen later the fatigue results from this batch were incompatible with the other results. The heat-treatment of 125 tons/in² material produced consistent results for all batches.

Microstructures in the quenched and tempered at 500° and 200° C conditions were typical of this class of steel given these treatments. Measurements by X-ray diffraction showed that in the latter condition the amount of retained austenite in the fatigue test-pieces was less than 5%.

Fatigue-testing Machines and Test-pieces

The fatigue-testing facilities have remained unchanged throughout the investigation; the testing machines are of the N.P.L. two-point loading, rotating-beam type described by Phillips.[9] The test-pieces have an unnotched, uniformly loaded, critical section and details of their design (identified as *F*) are given elsewhere.[10] A description of the fatigue-testing laboratory and equipment has also been given previously.[11]

Test-pieces from material in the 85 tons/in² condition were prepared by turning from $\frac{7}{8}$-in. dia. bar

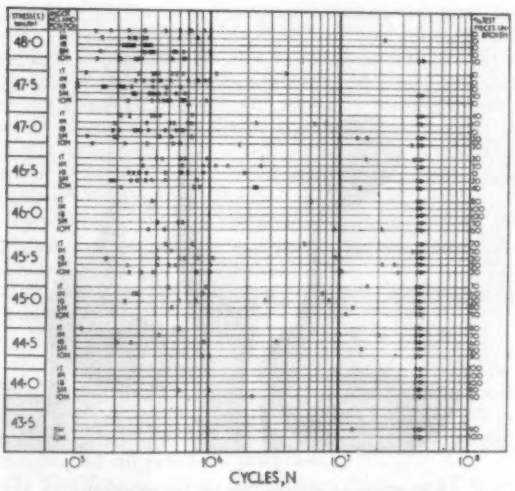


Fig. 3—Fatigue test results, 85 tons/in² condition (criterion for non-failures: 40 × 10⁶ cycles)

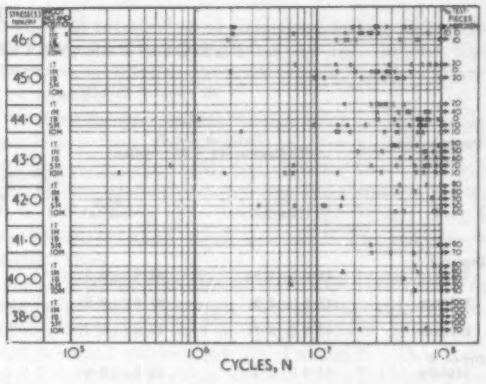


Fig. 4—Fatigue test results, 125 tons/in² condition (criterion for non-failures: 100 × 10⁶ cycles).

after heat-treatment and finally by hand-polishing longitudinally with various grades of metallographic polishing papers down to 000. In the case of the material in the 125 tons/in² condition, the method of test-piece preparation was modified. The initial blanks were rough machined to plus 0·030 in. of the final dimensions. They were then heat-treated, cylindrically ground to finished size, and finally longitudinally polished with diamond paste. The primary difference between the two methods of preparation was the substitution of grinding for turning.

A small series of fatigue tests at $\pm$ 55 tons/in² was completed on material heat-treated to 125 tons/in² to determine the effect of this change in preparation. The turned and polished specimens had a mean life of $3{\cdot}8 \times 10^6$ cycles whereas the ground and polished specimens had a mean life of $4{\cdot}3 \times 10^6$ cycles. This difference was not considered to be large enough to warrant rejection of the alternative method of preparation.

RESULTS OF FATIGUE EXPERIMENTS

Form of Results

The complete results for the two conditions of heat-treatment are shown in Figs. 3 and 4 respectively. According to the plan, for each condition ten tests were to be undertaken at each of the ten stress levels for each of the five ingot positions. For various reasons it was not possible to keep to this plan completely.

In the 85 tons/in² condition, one batch of 30 test-pieces from ingot 1 was damaged during machining and no attempt was made to test these. A further batch, as mentioned earlier, were found to be 'off-standard' and although they were tested (*see* stress level $\pm$ 46 tons/in²: ingot 1 in Fig. 3), the results were not included in the subsequent analysis which was, therefore, for ingot 1, restricted to eight of the intended ten stress levels. The full ten stress levels were completed for ingots 5 and 10.

The test-pieces in the 125 tons/in² condition required such a long duration of test that the time to complete the whole planned experiment became prohibitive. The programme was, therefore, shortened by increasing the stress increments and reducing the number of

Table V

FATIGUE PARAMETERS OF INGOTS 1, 5, AND 10

Position in Ingot	85 tons/in² Condition			125 tons/in² Condition		
	Fatigue Limit (Criterion 40 × 10⁶ cycles)		Fatigue Ratio	Fatigue Limit (Criterion 100 × 10⁶ cycles)		Fatigue Ratio
	95% ± tons/in²	50% ± tons/in²	F.L. (95%) / U.T.S.	95% ± tons/in²	50% ± tons/in²	F.L. (95%) / U.T.S.
Ingot 1 Top	44·0 (±2·5)	45·9 (±0·5)	0·511	39·1 (±2·3)	42·8 (±1·1)	0·317
Middle	43·8 (±2·5)	45·6 (±0·5)	0·510	39·7 (±2·0)	42·7 (±1·0)	0·321
Bottom	43·9 (±0·7)	45·0 (±0·4)	0·515	39·0 (±2·3)	42·7 (±1·1)	0·320
Ingot 5 Middle	42·6 (±2·2)	45·5 (±0·6)	0·495	39·2 (±2·0)	41·9 (±1·0)	0·314
Ingot 10 Middle	43·3 (±1·8)	45·8 (±0·6)	0·501	37·3 (±2·7)	41·4 (±1·2)	0·301

N.B.—The accuracy of each estimate of fatigue limit is shown by the figures in parenthesis. These figures ('t' × standard error) define a range in which the true fatigue limit will be expected to be with 95% confidence.

stress levels investigated to seven, in the case of ingot 1, and six each in the cases of ingots 5 and 10.

Analysis of Results

The definition of the fatigue limits has been described elsewhere[1, 12] together with a description of the necessary analysis.[1, 12, 13]

(i) The fatigue limit (95%) is that stress level at which 95% of tests can be expected to exceed 40×10^6 cycles for the 85 tons/in² condition and 100×10^6 cycles for the 125 tons/in² condition

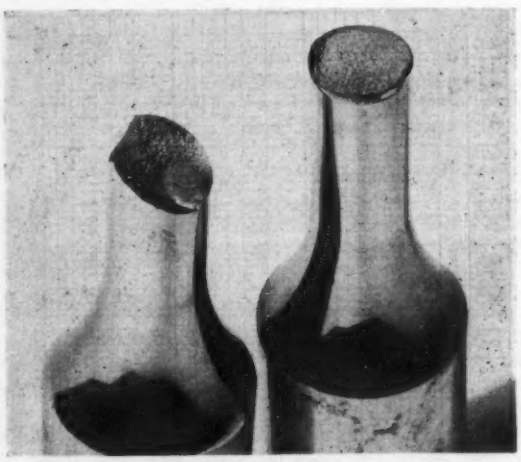
(ii) The fatigue limit (50%) is that stress level at which 50% of tests can be expected to exceed 40×10^6 cycles for the 85 tons/in² condition and 100×10^6 cycles for the 125 tons/in² condition.

The estimates of the fatigue limits for the separate ingot positions, and the errors associated with these estimates, are given in Table V. Clearly the fatigue properties are similar for all ingot positions whichever heat-treatment condition is considered. The difference between the largest and smallest fatigue limit (95%) observed is about 3% for the 85 tons/in² material and about 6% for the 125 tons/in² material; neither of these differences is considered of any practical importance.

A statistical analysis of the fatigue results produced different conclusions from those for the tensile tests. Generally both the tensile and fatigue strengths of material from the bottom of ingot 1 are relatively inferior to those of material from the top and middle of this ingot. But this cannot be stated as confidently for the fatigue results as for the tensile results; in fact only the fatigue limits (50%) for the 85 tons/in² material indicate any statistically significant differences between the positions within an ingot. Further, in direct contrast to the relative tensile strengths of the three ingots, the fatigue strength of material from the middle of ingot 1 is generally superior to that of material from the middle of ingots 5 and 10. In particular for the 125 tons/in² material, the fatigue limits (50% and 95%) of material from the middle of ingot 10 are statistically significantly lower than the corresponding fatigue limits for ingot 1.

A most important feature of the fatigue results is the marked falling away of the fatigue properties in the higher tensile condition. Table V shows that each fatigue limit estimate for the 85 tons/in² material is higher than the corresponding estimate for the 125 tons/in² material. This point is emphasized if the fatigue ratios are compared instead of the fatigue limits. To be strictly comparable the corresponding fatigue limit estimates should both be based on the same threshold value, which is not true in Table V; an estimate for the 125 tons/in² condition based on 40×10^6 cycles has little practical meaning. However, it is interesting to note that even the use of this estimate (overall fatigue limit (95%) ± 42·5 tons/in², fatigue ratio 0·344) demonstrates that the 85 tons/in² material (overall fatigue limit (95%) ± 43·5 tons/in², fatigue ratio 0·506) has superior fatigue properties.

The mean number of cycles to failure, at a given stress level, does not differ significantly between ingot positions for either heat-treatment condition. But the variation in the lives of test-pieces certainly



(*a*) 85 tons/in² condition (*b*) 125 tons/in² condition

Fig. 5—Typical examples of fractured test-pieces × 4

depends on the tensile strength of the material (*see* Figs. 3 and 4). Analysis of the results from two stress levels common to both experiments ($\pm$ 44·0 and 45·0 tons/in²) demonstrated that although 82% of the 100 tests made on the 85 tons/in² material exceeded 40×10^6 cycles against 55% for the 125 tons/in² material, nevertheless 13% of the 85 tons/in² test-pieces fractured before 4×10^6 cycles against only 3·75% of the 125 tons/in² test-pieces.

Examination of Fractures

The fractured surface of all broken test-pieces was examined under a stereoscopic microscope at a magnification of × 100 immediately after failure. A proportion were also examined at higher power. In no case, in either condition of heat-treatment, could the initiation of fracture be directly attributed to a non-metallic inclusion, although in some instances, particularly samples from ingot 10 in the high-strength condition, rounded cavities could be detected near the origin of fracture which might have been the site of an inclusion. Many of the test-pieces heat-treated to 125 tons/in² tensile strength exhibited ' white spots ' in the fracture, very much as reported by Frith.[2] These white spots occurred more frequently at the lower stress levels and their size invariably fell within the range 0·004–0·15 in. diameter. The most careful examination of these white spots did not reveal the presence of a non-metallic inclusion. The spots did not occur when failure could be attributed directly to the presence of a surface imperfection or when failure took place in the fillet radii. It is suggested that the incidence of these white spots is influenced more by the state of the surface stresses than by the presence of non-metallic inclusions. Support for this idea is provided by the behaviour of nitrided specimens which generally exhibit a white spot at the point of initiation of the sub-surface failure when fractured in fatigue.

The shape of the fractured surfaces in the 85 tons/in² material was rather unusual. From the point of initiation the fractures appeared to have progressed normal to the surface for some distance and then to have changed direction so that the remainder of the fracture was at 45° to the original direction. The portion of fracture surface normal to the specimen surface was found to be in the form of a round plateau, usually very smooth, whilst the remainder was of a form more usually associated with a torsional failure. This type of fracture occurred in all cases for material in the 85 tons/in² condition.

The fractures in the 125 tons/in² material were all virtually normal to the specimen surface and comparatively smooth in appearance.

Examples of typical fractures in both conditions of heat-treatment are shown in Fig. 5.

CONCLUSIONS

(1) The fatigue properties of this cast of En 24 steel have been shown to be uniform throughout when heat-treated to either 85 or 125 tons/in² tensile strength, for all practical purposes.

(2) The fatigue limits (95%) vary from $\pm$ 42·6 to $\pm$ 44·0 tons/in² for the 85 tons/in² steel (based on 40×10^6 cycles) and from $\pm$ 37·3 to $\pm$ 39·7 tons/in² for the 125 tons/in² steel (based on 100×10^6 cycles). The corresponding variations for the fatigue limits (50%) are $\pm$ 45·0 to $\pm$ 45·9 tons/in² and $\pm$ 41·4 to 42·8 tons/in².

(3) The fatigue ratios, based on the 95% fatigue limit, vary from 0·495 to 0·515 for the 85 tons/in² steel but only from 0·301 to 0·331 for the 125 tons/in² steel. Even when the same threshold criterion is used (40×10^6 cycles) the overall fatigue ratio is 0·506 for the 85 tons/in² steel against 0·344 for the 125 tons/in² steel.

(4) This particular cast of steel was a comparatively clean one and although the distribution of non-metallic inclusions was fairly uniform it did not conform to that postulated by Palazzi.[8] The top of ingot 1 contained rather more silicate inclusions than the other portions but they were generally very small. Considering all the ingot positions collectively the largest inclusions were predominantly silicates. No direct association between the inclusion content of the steel and fatigue failure could be detected. There was some indication that certain inclusions may have contributed to the initiation of a few failures of material in the 125 tons/in² condition.

Acknowledgments

The authors wish to thank the former Chairman, Dr. H. H. Burton, the present Chairman, Mr. J. E. Russell, and past and present members of the Fatigue and Inclusions Committee for their encouraging and helpful discussion on this work. They are also indebted to Dr. M. L. Becker (formerly Head of the Metallurgy Division of B.I.S.R.A.) for his advice in the course of the work and to Mr. T. R. Allmand who undertook much of the experimental work, particularly the metallographic work. Mr. H. Scott supervised the running of the fatigue machines and the authors would like to record their appreciation of the skill and care he devoted to it.

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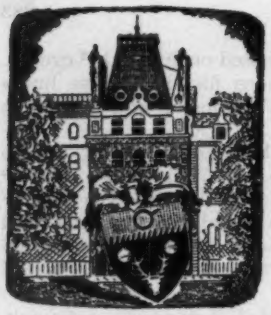


PROCEEDINGS OF THE INSTITUTE

Reports of Council · Meetings · Notices

Annual General Meeting, 1958

The EIGHTY-NINTH ANNUAL GENERAL MEETING of THE IRON AND STEEL INSTITUTE was held on Wednesday and Thursday, 7th and 8th May, 1958, at the Offices of the Institute, 4 Grosvenor Gardens, London, S.W.1. Mr. A. H. INGEN-HOUSZ, the Retiring President, was in the Chair at the beginning of the Meeting, his place being taken later by Mr. C. R. WHEELER, C.B.E., the new President.

Discussions at some of the technical sessions are given below; further discussions will be published in subsequent issues of the *Journal*.

JOINT DISCUSSION ON STEELS CONTAINING CHROMIUM, VANADIUM, AND MOLYBDENUM

This discussion was based on the following papers (the dates of publication in the *Journal* are given in parentheses):

"**Carbide Precipitation in Several Steels containing Chromium and Vanadium,**" *by A. K. Seal and R. W. K. Honeycombe* (1958, vol. 188, Jan., pp. 9–15).

"**The Effect of Tantalum and Niobium on the Tempering of Certain Vanadium and Molybdenum Steels,**" *by A. K. Seal and R. W. K. Honeycombe* (1958, vol. 188, Apr., pp. 343–350).

"**The Tempering of Low-alloy Creep-resistant Steels containing Chromium, Molybdenum, and Vanadium,**" *by E. Smith and J. Nutting* (1957, vol. 187, Dec., pp. 314–329).

Professor **R. W. K. Honeycombe** (Sheffield University), who presented the first two papers, said: We are discussing the results of some of the work carried out over the last five years on the precipitation of carbides in steels. Much progress has been made, primarily by the application of electron microscopic methods to the study of these carbides. But, substantial though the progress has been, most of the work which is to be discussed has been done with electron microscopes with resolution between 30 and 50 Å. There are now in operation instruments which can resolve less than 10 Å, and which can do better than this in certain circumstances. It seems reasonable to prophesy, therefore, that the next ten years will see even greater progress than the immediate post-war era.

Added to this, the use of electronic methods in general is becoming very extensive indeed. The electron microscope is not the only means at our disposal, and indeed without associated electron diffraction techniques the results of electron microscopy would be inadequate. We can now add the electron probe analyser, which is becoming fairly well known. I think that there is a great future for such instruments, by which it is possible, by using a fine electron beam of about 1 μ in diameter, to analyse by X-ray fluorescence the very small area involved, the amount of material in these analyses being incredibly small.

The latest development, at the Cavendish Laboratory, is that of a scanning electron probe analyser, where an actual electron image, as in the electron microscope, is formed by a scanning electron spot of very small dimensions. It is possible to stop the spot at any point of interest and then analyse that very small region by means of the X-rays emitted from it when the electron beam impinges on it.

These pieces of equipment may seem complicated, but I am sure that we can rely on the electrical firms to develop these instruments to the point when we will be able to use them for research and even routine examinations. The electron microscope has certainly reached this stage, and I am sure that these other and more exotic instruments will soon do so.

I say this to emphasize the fact that we are at a half-way stage, and I believe that progress will become much more rapid in the next few years. I only hope that those laboratories which have started this work will have the opportunity to continue it. The equipment is extremely expensive, and many of us have great difficulties in getting the necessary money, but I feel that the electron microscope has taught a valuable lesson, and that in future there will be more encouragement to use these modern physical methods.

Dr. **J. Nutting** (Cambridge University), in presenting the third paper, said: Five years ago electron microscopes were available with resolving powers of 20–30 Å, but we had no satisfactory techniques which would allow us to resolve 20–30 Å, and one of the first things to be done was to develop a method for preparing replicas from alloy steels. This was done by Dr. Smith. He first of all used a method which had been described by Bradley, of taking plastic replicas and evaporating carbon on to them, but he discovered that this was not very satisfactory, and this led him to develop the direct carbon replica process, which has been successfully used by ourselves and other workers.

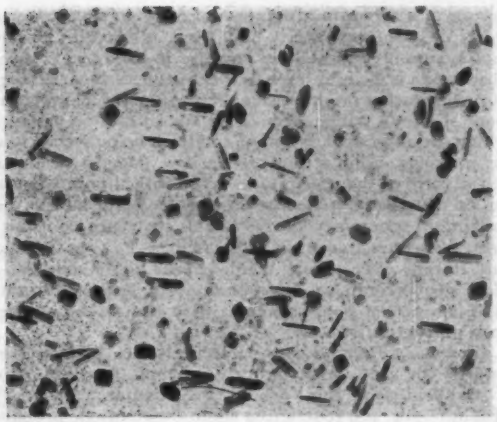


Fig. A—Widmanstätten pattern of V_4C_3 particles after tempering at 650° C

Mr. **F. B. Pickering** (United Steel Companies Ltd.): I congratulate the authors on the excellence of both their results and their electron micrographs. At the United Steel Companies we have been making a systematic study of the effect of molybdenum, vanadium, tungsten, and chromium in low-alloy steels containing boron, and also the effects of various alloy additions to 12%Cr steels and to still higher chromium steels. We have accumulated much data involving not only hardness measurements but also electron and X-ray diffraction analysis of the particles, and electron microscopy.

I shall divide my remarks into three main topics: (*a*) the precipitations in vanadium steels; (*b*) the precipitations in molybdenum steels; and (*c*) the changes in high-chromium steels.

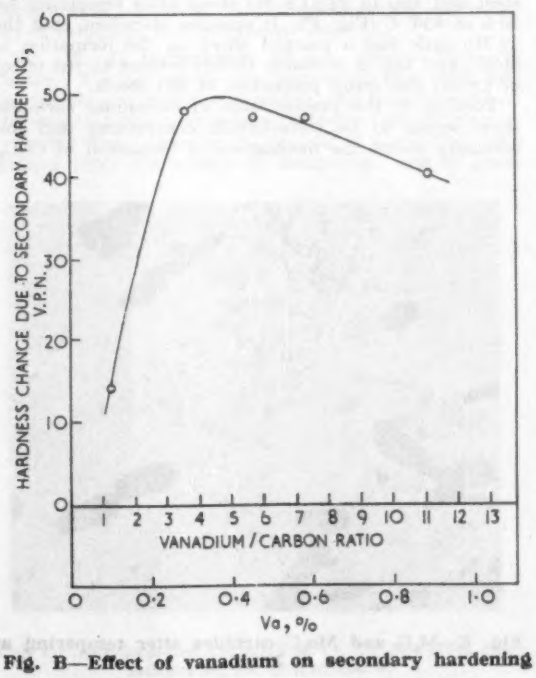


Fig. B—Effect of vanadium on secondary hardening

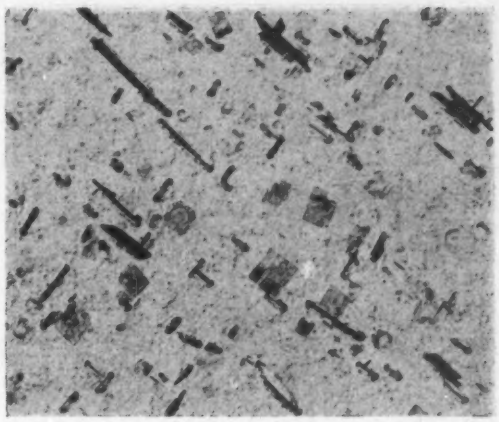


Fig. C—Vanadium carbide plates in delta ferrite

Before dealing with these three topics, however, I should like to discuss the initial structures. We have never identified ϵ-carbide as such, probably because our techniques have not been sensitive enough, or we have not tempered the steels at low enough temperatures. I should like to ask Dr. Nutting whether there are any morphological differences between the two forms, ϵ and Fe_3C, which could aid identification.

Turning to the precipitation of vanadium carbide, we have examined the effect of increasing vanadium content in ½%Mo–B steels. In the base ½%Mo steel, Mo_2C was the precipitating carbide, but as little as 0·1%V changed the carbide type to V_4C_3 and introduced marked secondary hardening. At this low vanadium content, however, $M_{23}C_6$ (the remnant influence of the molybdenum present) was still observed at the higher tempering temperatures. Above 0·2%V, however, the only alloy carbide observed was V_4C_3.

We concur completely with the results of Smith and Nutting on the secondary-hardening processes in vanadium steels, namely, that the V_4C_3 precipitates by separate nucleation and the Fe_3C redissolves, until there is a fine dense precipitate of V_4C_3 occurring throughout the steel. During the over-ageing we observed rather more marked Widmanstätten patterns of V_4C_3 carbides than that shown by Smith and Nutting, Fig. A, and we observed also similar electron diffraction effects to those noted by Smith and Nutting, together with marked preferred orientation maxima on the rings from the Widmanstätten precipitate. The authors' interpretation of the particle shape in the early stages of formation of V_4C_3 agrees with our views on the growth of the V_4C_3 plates which, in general, maintain this shape but lose coherency as they grow, and this results in the over-ageing or softening.

Dr. Nutting suggests that molybdenum does not enter into solid solution in V_4C_3, but this is contrary to our experience. We find that after tempering for 100 h at 700° C the lattice parameters of the V_4C_3 decrease as the vanadium content increases. On the basis of the results of Krainer, these parameter changes indicate that up to 40 at.-%Mo can enter the V_4C_3 in the 0·1%V steel, but as the vanadium content increases the proportion of Mo/V in the V_4C_3 decreases. In view of the effects observed by Seal and Honeycombe, I should like to ask whether molybdenum dissolves in Cr_7C_3.

Also in agreement with Smith and Nutting, we find that an increase in vanadium content first retards the secondary-hardening reaction up to V/C ratio of 3:1, but that

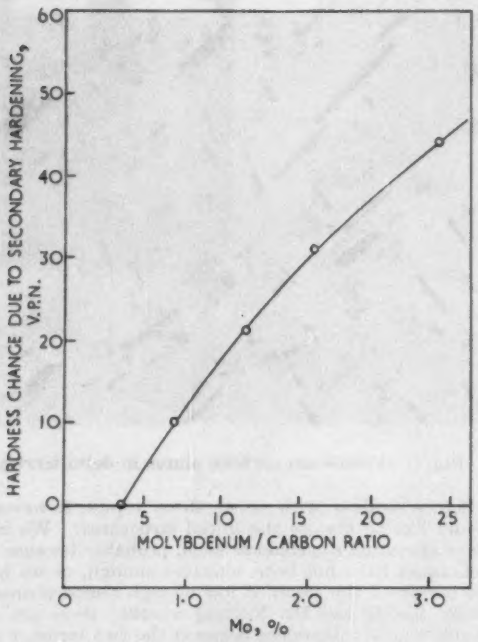


Fig. D—Effect of molybdenum on secondary hardening

further increases in vanadium content then accelerate the reaction. The reason for the initial retardation could be either stabilization of the Fe_3C by vanadium or possibly the precipitation occurring over longer times with increasing vanadium content, while the acceleration at higher V/C ratios is probably due to increased supersaturation of V_4C_3 causing more rapid precipitation and over-ageing. Increasing the V/C ratio also produces an optimum in the intensity of the secondary hardening at a V/C ratio of about 5:1 (Fig. *B*).

From the microstructures it appears that the increase in secondary hardening with increasing vanadium is due to an increased number of V_4C_3 precipitates, whilst the decrease in hardening above the optimum vanadium content is the result of a greater tendency for the V_4C_3 to coarsen more rapidly as the V/C ratio increases.

With regard to the recrystallization effects observed by Smith and Nutting, much of our work supports many of the points put forward by them, but we would also point out that recrystallization can occur in the presence of V_4C_3 at martensite plate boundaries under the activation of creep deformation. The initial structure of a 1%Cr–Mo–V steel in the tempered condition, 9 h at 700° C, consisted of a fine dispersion of V_4C_3 with larger V_4C_3 particles at the acicular ferrite grain boundaries. This structure was retained after creep testing to 0·15% strain in 5000 h at 565° C, but after creep testing up to the same strain and at the same temperature in 11,000 h considerable recrystallization occurred.

Turning to the effect of vanadium in the high-chromium steels, we have been mainly interested in the 12–13%Cr steels, and we are in agreement with Seal and Honeycombe in finding no V_4C_3 as a separate phase (even in 12%Cr steel with 1%V), if there is no δ-ferrite present. If δ-ferrite is present in the steel we find that V_4C_3 precipitates within the ferrite, and Fig. *C* shows the structures in which we can definitely see small, thin, rectangular plates.

In Mo–B steels, the evidence we have obtained is all in favour of a separate nucleation of Mo_2C in the same way that V_4C_3 is separately nucleated, although at low-molybdenum contents the increase in hardness is either absent or very slight. We find that the very early stages of precipitation produce a fine dense cloud-like precipitation of Mo_2C which gives rise to diffraction patterns which are imperfect: some of the lines being either weak or missing. The strong reflections suggest that the initial form of the precipitates is that of thin plates parallel to the basal plane of the hexagonal lattice. At maximum secondary hardness the structure contains a very dense precipitate of Mo_2C, whilst Fe_3C redissolves, but during over-ageing what appear to be needles of Mo_2C grow from this dense precipitate, and we obtain a Widmanstätten pattern again, with the attendant electron diffraction effects.

Increasing molybdenum content to 2% progressively retards the rate of secondary hardening, possibly owing to stablization of the Fe_3C by molybdenum. Above 2% Mo, however, the reaction is again accelerated, probably because of increased over-ageing due to the increased degree of supersaturation, although the effect of lower transformation temperatures cannot be excluded. The effect of molybdenum contents up to 3% is progressively to increase the degree of secondary hardening (Fig. *D*). The microstructures indicate that this is due to increased numbers of precipitated Mo_2C particles.

The over-ageing of the molybdenum steels is not as straightforward as that of the vanadium steels, because between 1% and 2%Mo, two transitional carbides, $M_{23}C_6$ and M_6C_b, tend to appear during the approach to the final equilibrium carbide M_6C. These two carbides form most readily in the 1¼%Mo steel, but are generally absent in the 3%Mo steel, when the over-ageing reaction is one from Mo_2C to M_6C *in situ* mainly at grain boundaries. As the M_6C grows so the Mo_2C needles tend to coarsen and re-dissolve (Fig. *E*).

Smith and Nutting state that no Mo_2C was observed in the 3%Cr–Mo steel, and yet we find extensive Mo_2C formation with both Fe_3C and Cr_7C_3 in a 1%Cr–Mo steel and also in 2¼%Cr–Mo steels after tempering for 30 h at 650° C (Fig. *F*). It appears, therefore, that the Cr/Mo ratio has a marked effect on the formation of Mo_2C, and this is probably closely related to the effect of Cr on the creep properties of Mo steels.

Turning to the precipitation of chromium carbides, there seems to be considerable controversy and uncertainty about the mechanism of formation of Cr_7C_3.

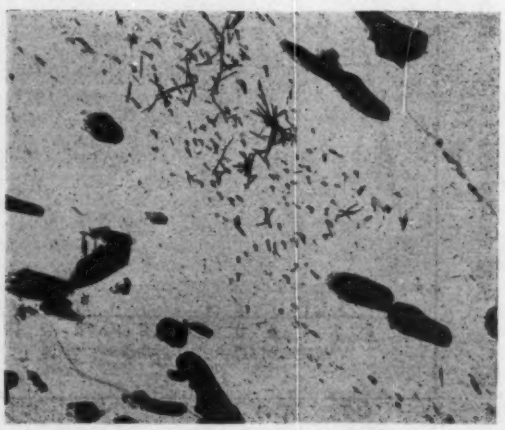


Fig. E—M_6C and Mo_2C carbides after tempering at 700° C

Those who work with high-chromium steels postulate that a separate nucleation effect is observed, while those who work with low-chromium steels take a different view. With regard to high-chromium steels, we find that in 12%Cr steel containing 0·1%C the formation of Cr_7C_3 gives rise to a secondary-hardening type of reaction, the dense precipitate of Cr_7C_3 being separately nucleated. A further increase in tempering produces a series of very fine plates in the matrix of the specimen which are Cr_7C_3 (Fig. *G*), whereas the remnant Fe_3C plates, which are bigger, are developing a dot–dash effect, or fragmentation, which is probably the same effect seen by Seal and Honeycombe in some of their photographs. Our work supports the belief that Cr_7C_3 forms by a separate nucleation, although the *in situ* transformation of $Fe_3C \rightarrow Cr_7C_3$ is also possible in view of the fragmentation effects in the Fe_3C. On the other hand, all the evidence we possess tends to show that in a 4%Cr steel, the change from $Fe_3C \rightarrow Cr_7C_3$ occurs *in situ* (Fig. *H*).

The secondary hardening by Cr_7C_3 is not large in plain 12%Cr steels, but we agree with Seal and Honeycombe that Mo and V can increase the intensity of this secondary hardening; ½% Mo increases the hardness by 35 D.P.N. and ½%V by 25 D.P.N. We have found no effect of Nb, but this is probably due to the use of a too low solution treatment temperature.

Finally, I note that Dr. Nutting has pointed out some of the fallacies inherent in the use of tempering parameters but I must say a few words in their defence. It is true that their use cannot adequately represent a precipitation reaction, nor do they show changes in structure, such as grain size, at the same hardness value. Consequently, they are not suitable for investigating thoroughly the changes in structure occurring during tempering; isothermal or isochronal curves are preferable

Nevertheless, for use in rapid heat-treatment, and for enabling much information to be presented on one diagram, tempering parameter curves possess a real advantage. Also, they are more than useful in enabling rapid sorting of steels, by relatively few tests, into categories of increasing tempering resistance, and hence are of value also in development work.

Dr. **H. J. Goldschmidt** (B.S.A. Group Research Centre): I congratulate the authors on this work and on the elegant manner in which they have brought all the various techniques to bear on the carbide problem.

In the first paper the re-solution of cementite, once formed from martensite on tempering, was of great interest. On tempering at the highest temperature used, 700° C, does the Cr_7C_3 carbide form immediately, without prior cementite, or does the sequence of events occurring at lower temperatures (namely, formation, re-solution of Fe_3C, formation of Cr_7C_3), apply in principle to all sub-transformation temperatures, only at an accelerated rate at the higher ones? It is tempting to follow the work through to shorter times and to enquire whether, after a few minutes at 700° C, the cementite stage would still appear, with the same hardness sequence, as well as to follow the pre-precipitation mechanism from Guiner–Preston zone effects in single crystals (if these can be produced).

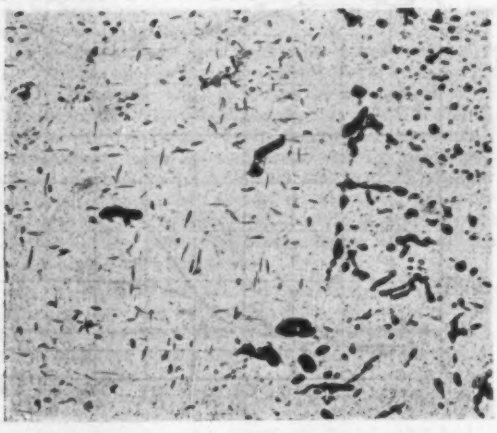


Fig. F—Molybdenum carbides in 1% Cr–Mo steel after tempering for 30 h at 650° C

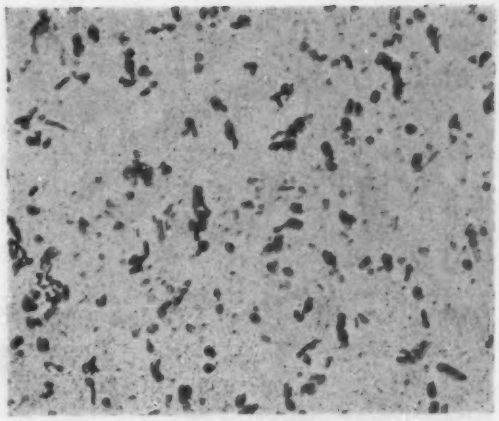


Fig. G—Fine plates of Cr_7C_3 in 12%Cr steel tempered at 500° C for 240 h

Would the authors conclude that the chromium carbide formation and the cementite solution proceed as two independent reactions, or are they not at least partly interrelated, in so far as one nucleates preferentially at points abandoned by the other? Figure 8 seems to support the latter, but the weight of the authors' results the former.

It seems to be generally the case, also in the Ta, Nb, and Mo steels, that, whatever the carbide-forming alloy addition, cementite forms first. This interesting finding seems to put the iron carbide into quite a different category to the rest, as of course structurally it is. I wonder whether the solution of cementite and the simultaneous nucleation of alloy carbide, should be regarded as quite separate phenomena, or whether they are partially in a 'cause-and-effect' relation, in so far as cementite must re-enter into solution, to replenish the parent lattice with the carbon needed to precipitate the alloy carbide.

A further question of interest is that of the chromium content of the initial cementite $(Fe,Cr)_3C$, and whether this is already similar to that of the subsequent $(Cr,Fe)_7C_3$ carbide, which itself is an iron-saturated solution. It is possible that the highly dispersed cementite is already quite high in chromium. A simple analysis of the extracts may perhaps decide that.

Furthermore, were there any observable changes in lattice spacing or in line width of Fe_3C and Cr_7C_3 during the course of precipitation? These would be significant.

Figure *I* shows a set of our results on low Cr–Mo steels which would tie in with the authors' first stages of nucleation. It illustrates changes of the cementite lattice dimensions in a series of 0·3%C steels with varying molybdenum content, and at several chromium levels. In cementite, whilst the Cr atom readily replaces iron, the Mo atom also enters, but only metastably.

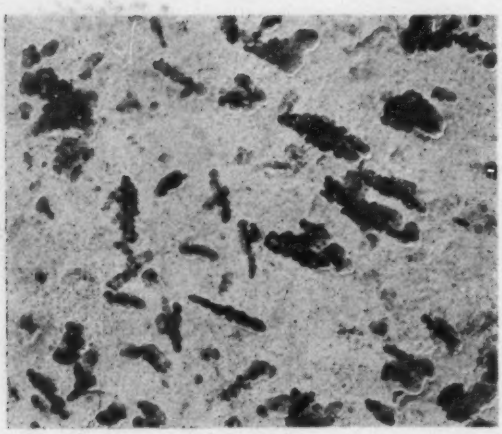


Fig. H—*In situ* transformation of Fe_3C to Cr_7C_3 in 4%Cr steel tempered at 450° C

It is in fact too large to remain in stable solution, but is present in supersaturation, producing considerable lattice strains. These are shown by line broadening, the line width increasing with rising Mo contents, from perfectly sharp lines without molybdenum. The expansion is not isotropic, and Fig. *I* shows how the three axes of the orthorhomic unit cell change with Mo-content: The *a* axis remains fairly constant, the *b* axis attains a minimum, which coincides with a maximum in the *c* axis. It would appear that the Mo atom when it enters the Fe_3C lattice unstably, tends to create already a carbon configuration around itself, corresponding to the formation of incipient Mo-richer carbide: in the authors' terms of Mo_2C.

One effect is that, within the cementite lattice, the carbon content will vary from one cementite unit cell to the other, depending on whether an Mo atom is near or not. Such heterogeneity on a lattice scale would again imply stress and increasing line broadening.

The point of principle I want to make is that such carbide → carbide precipitation can be, I think, almost as important as matrix → carbide precipitation, and that it should receive more attention, perhaps by single crystal as well as powder work.

Passing on to the authors' vanadium steels, one point that was not mentioned is the liability of defect lattice formation in vanadium carbide, which I think can play an important part. Should the formula 'V_4C_3' here be taken as the strictly correct one, or is it only nominal, and might it still be 'VC,' with perhaps some vacant carbon sites? This difference becomes the more significant if we have Nb and Ta also present, as their carbides would definitely favour the higher carbon contents. Also, we know that V_4C_3 is isomorphous with carbon-saturated austenite, and that, historically, it will have originated from austenite, though with the intervening martensitic shear. I should like to ask whether there are any signs of orientation and coherency between the matrix crystals and V_4C_3 which might reflect any such origin. Moreover, in the vanadium–carbon system, the cubic carbide is not the only one, but a hexagonal carbide V_2C is also known to exist. It would, I suggest, be extremely interesting to see whether there is any such intermediate stage during the precipitation of V_4C_3. May I suggest that, before decomposition of austenite, stacking faults would occur which would give within the face-centred cubic lattice, localized pseudo-hexagonal regions, structurally related as the two vanadium carbides V_2C and V_4C_3. This also ties up with the fact that the f.c. cube can accommodate more carbon than the c.p. hexagonal lattice. In fact, I cannot help wondering whether, around the sites of the vanadium–carbide crystallite there would not exist in the matrix a very thin film of retained austenite, as is the case at the normal grain boundaries. Such coherency between V_4C_3 and original austenite is worth following up.

One small factual question: in steel *A*, only Fe_3C, Cr_7C_3, and $Cr_{23}C_6$ are mentioned as the carbides, although it contained 0·38%V. Is it implied that this amount of vanadium remains in solid solution in the matrix or in these other carbides, and from what percentage upwards would the authors say that vanadium exerts its individuality and produces its own carbide V_4C_3?

Referring to the second paper, is it again implied that 'NbC' is the correct composition for niobium carbide? It has been stated as 'Nb_4C_3' (analogous to V_4C_3), and here again defect lattice effects may well become important.

I am not clear whether we should conclude that the niobium and tantalum carbide precipitate as one solid solution with the vanadium carbide, as suggested in the text, or whether V_4C_3 and (Nb,Ta)C are two separate phases as the X-ray spacings in Table III seem to indicate.

In the molybdenum steels, the concept of Mo_2C as a non-equilibrium condition towards eventual η-carbide (M_6C) formation seems particularly interesting, and agrees with our results. I wonder however, whether, in the steel, the η-carbide should be considered as a reaction product between the Mo_2C and Fe_3C, formed previously, or as a fresh precipitate from the matrix, whilst the former have gone into solution. There is perhaps one further complication not mentioned, in that, for Mo and V steels, the $M_{23}C_6$ carbide is liable to form, besides M_6C, particularly for higher Cr contents.

Dr. R. G. Baker (Cambridge University): The presence of ϵ-carbide in low-carbon, low-alloy steels has been

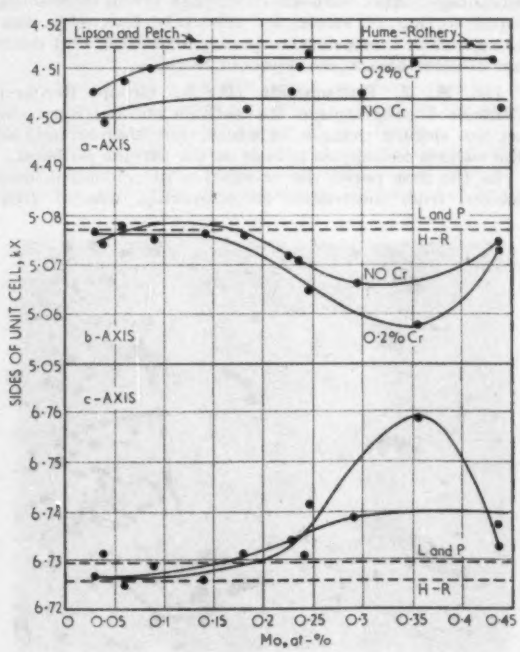


Fig. I—Change in lattice dimensions of cementite with varying chromium and molybdenum contents

deduced from a study of the carbides in extraction replicas. The techniques used have included electron diffraction, X-ray fluorescence analysis, X-ray diffraction, and electron microscopy. When the results obtained from these techniques are considered together, there is very good evidence for the presence of ϵ-carbide.

We have studied the tempering of a Ni–Cr, a Ni–Cr–Mo, and a Cr–Mo steel,* each containing 0·2% carbon, and ϵ-carbide was detected after tempering at 400° and 500° C in all three steels. The electron diffraction patterns show very diffuse rings due to ϵ-carbide after tempering for short times at 400° C. Further tempering results in the appearance of spotty cementite rings which slowly become more complete. However, by itself this evidence is insufficient since it rests on the measurement of diffuse and spotty rings. X-ray fluoresence analysis shows that the total amount of carbide changes on tempering, increasing by a factor of about five between ½ and 100 hours at 400° C. Thus, after ½ h at 400° C the ϵ-carbide must be in metastable equilibrium with about 0·15% carbon in solution. The chromium and manganese contents increase at first during tempering, but reach a maximum after about 50 h at 400° C (or 2 h at 500° C) and then decrease. It is difficult to explain these results except on the assumption that ϵ-carbide is transforming to cementite. The initial increase in alloy content is thought to be due to replacement by Cr and Mn of the Fe in the ϵ-carbide. The subsequent decrease occurs when ϵ-carbide is being replaced by cementite, provided that the solubility of Cr and Mn in cementite is less than in ϵ-carbide at the same temperature.

The morphological changes observed in the electron microscope support the view that ϵ-carbide is changing to cementite on tempering at 400° and 500° C. The carbide plates become gradually thicker and better defined as the transformation proceeds. This is in close agreement with the experimental observations of Crafts and Lamont† in their study of the transformation of ϵ-carbide to cementite in low-vanadium steels.

Smith and Nutting found that, on tempering their V–C steel at low temperatures, polygonal grains were formed, whilst at higher temperatures the acicular grains persisted. If particles of carbide are present in the matrix in which the trains are tending to grow, they will interact with the grain boundaries and tend to obstruct the growth of the grains. It can be shown that it is necessary to exert a force on such a boundary to get it away from the precipitated particle. With a particle of radius r and where the interfacial tension is λ, the force has to be proportional to $r \times \lambda$ to tear away the boundary from the precipitate.

In the case of the steel tempered at low temperatures the carbides are very small and coherent; thus both the interfacial tension and the radius are very small, and the force needed to tear the boundary away from the particle is also small. In the case of the steel tempered at a higher temperature, larger non-coherent carbide particles are formed much more quickly at the ferrite grain boundaries and these exert a much greater restraint to grain growth. This may explain why recrystallization or grain growth is observed at low tempering temperatures in this steel and not at higher temperatures.

Dr. K. W. Andrews (United Steel Companies Limited): I should also like to congratulate the authors on the excellent work they have done in providing us with so much evidence about the early stages of tempering in low-alloy steels.

My colleague Dr. Hughes and I have been interested in

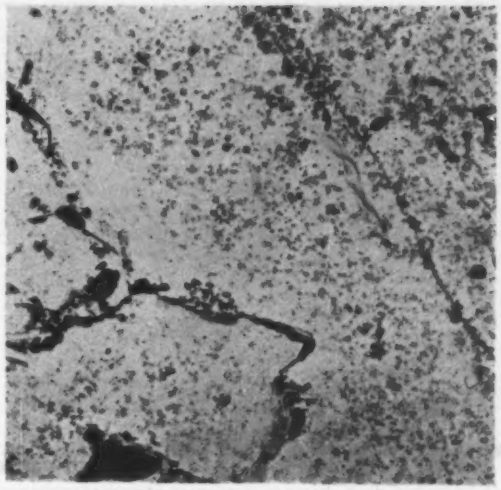


Fig. J—Carbon extraction replica from 0·1% C–0·6% Mo–0·25%V steel, normalized and tempered for 250 h at 700° C. Deep initial etch × 20,000

carbide precipitation and reaction processes for some time.* Our main instrument has been X-ray diffraction —often of residues separated by electrolysis.† As the authors indicate, there is some doubt as to the smallest particles which can be satisfactorily separated by such techniques. The electron microscope undoubtedly has the advantage for the study of the early stages of tempering and carbide reaction sequences particularly at lower temperatures. Nevertheless there is a considerable region where the two techniques overlap. Given the right choice of conditions, it is possible to use electrolytic separation followed by X-ray analysis towards the lower extreme. We do not think the 5% HCl solution used by Seal and Honeycombe is the most suitable for the types of steel investigated. For 9%Cr steels we would have used ferric chloride solution and for the low-alloy steels a sodium citrate–potassium thiocyanate solution. With the latter solution we succeeded in isolating ϵ-carbide in a steel containing 0·5%Cr and 0·4%C. We have recovered fine particles of Mo_2C when other electrolytes have dissolved them. The recovery of Mo_2C in 10 h at 550° C is perhaps comparable with Cr_7C_3 in 1 h at 500° C or V_4C_3 in 1 h at 550° C—the earliest times at which Seal and Honeycombe found these phases. I should, however, agree that reliance cannot be placed on the above X-ray results in this region and the results from various techniques must be combined. One disadvantage of electron diffraction should be noted: the diffraction patterns do not lend themselves to the same degree of accuracy in interpretation or measurement as X-ray patterns. If there are missing lines due to plate or rod formation, or patterns are superimposed, confusion is likely to arise. Such effects are easier to sort out with X-ray patterns.

There are a few other points of detail on which I should like to comment. One is the solubility of Mo in V_4C_3; Smith and Nutting seem to doubt whether this occurs. As Mr. Pickering has stated, we have definite evidence that Mo can in fact replace V in V_4C_3. This derives (*a*) from a correlation between the lattice parameter of the carbide and the V and Mo contents of the steel, and (*b*) from X-ray fluorescent analysis of the carbide extracted

* R. G. Baker and J. Nutting: To be published.

† W. Crafts and J. L. Lamont: *Trans. A.I.M.E.*, 1949, vol. 180, p. 471.

* K. W. Andrews and H. Hughes: *Iron Steel*, 1958, vol. 31, No. 2, Feb., pp. 43–50.

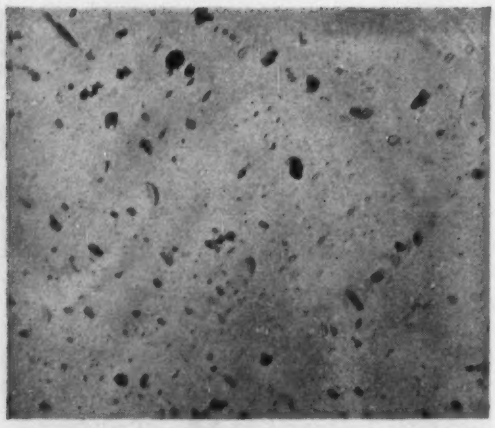


Fig. K—Same as Fig. J, but with light initial etch × 20,000

from a steel with about ½%Mo and 1%V. As Dr. Goldschmidt has said, the possibility of nitrogen entering into this phase must not be overlooked. One might also even add oxygen and vacancies (other than those already present at the V_4C_3 composition). Seal and Honeycombe in their second paper suggest that TaC and NbC dissolve in this phase and reduce the growth rate of the particles, and prior to that have increased the maximum hardness due to coherency strains. This is an interesting principle and deserves further attention. In a more general way it is known that a number of carbide formers together usually produce a stronger steel than any of them separately (cf. the effects described by Harris and Child in a "A Symposium on High-temperature Steels and Alloys for Gas Turbines," *Iron Steel Inst. Spec. Rep.* No. 43, pp. 67–80: 1952). Another topic in connection with these papers is the use of the tempering parameter $T(20 + \log t)$ which is discussed by Drs. Smith and Nutting. This is sometimes stated to be derived from reaction rate theory although it seems to conform only in part to this theory when applied to tempering. When applied to creep, the conditions are slightly different and it would seem to follow directly from reaction rate theory. In the tempering case, the plot of hardness against parameter, i.e. the tempering curve, may broaden out into a band. In fact, if simple reaction rate theory is obeyed, the single curve $T(C + \log t)$/hardness will only be followed for points at one temperature. In fact *for a given hardness* $T(C + \log t) = B + BT$, where, at that hardness level, B is a constant and D another constant which approaches zero in the case where a single curve appears. This relationship would account for the points shown in Fig. 24 by Smith and Nutting and, for example, if the 650° C points are 'corrected' to bring them into the 600° C curve the same constants would bring the other points on. There is, however, still some doubt as to the precise meaning of the tempering parameter and how it should be used. In spite of this, we have found the parameter useful. Thus, in one series of steels it brought out relationships which would have needed a great deal more data to establish. The parameter values for the disappearance of Fe_3C and for the appearance of M_6C respectively decreased in a regular manner as the Mo content increased.

Finally I should like to refer to the occurrence of hexagonal carbide phases such as ϵ–Fe_3C and Mo_2C. These two may be regarded as being near the lower end and upper end respectively of a whole range of intermetallic phases of the general formula M_2X where M is the metal atom and X the interstitial, C or N. Various members of this series and solid solutions can occur in a variety of steels. The V_2C carbide mentioned by Dr. Goldschmidt is one member of this class of compounds. They are based on a h.c.p. arrangement of metal ions (with interstitial C or N). Although the 2 : 1 ratio is often found, the ratio of metal atoms to non-metal may depart from it in some cases. The cubic carbides and nitrides are analogous, being based on the corresponding cubic close-packing of metal atoms. One might therefore expect ϵ-carbide to be stabilized by elements such as Cr, Mo, and V, although it does not follow that complete solid solutions exist. It seemed that the molybdenum-rich carbide also appears later on at higher temperatures whereas the ϵ-carbide has been replaced by Fe_3C in some steels. It would be interesting to know whether, in the complete absence of Cr or other alloy elements, the direct enrichment of ϵ-carbide by Mo can be brought about, so that the formation of cementite is prevented and hardening due to ϵ-carbide retained as long as possible.

Mr. F. G. Haynes (Mond Nickel Co., Ltd.): I should like to describe some work at Metropolitan-Vickers, with which I was formerly associated, where a 0·1%C–0·6% Mo–0·25%V steel was examined in the tempered condition after normalizing. Initially the methods which Smith and Nutting describe were used to prepare carbon extraction replicas, and Fig. *J* shows the sort of results obtained: a cloud of precipitate of no regular orientation.

We felt very dissatisfied with this, feeling strongly that there should be a regular pattern in the precipitate, and so we started to investigate the method of preparing these carbon extraction replicas. It was postulated that the initial etch was perhaps much too deep, so that fine carbides were being released by the action of that etch and were lying about loosely on the surface and not being removed by the cleaning techniques applied. In common with other workers, we put a Formvar film on and stripped it off before applying the carbon film, in the hope that this removed loose material; but by examining a succession of these cleaning replicas we found that we had loose material left even after several such cleaning procedures.

We did a series of experiments with very short etching times and from these arrived at the use of 0·1% nital for times of the order of 5–15 s. Figure *K* shows the same specimen prepared in this way, and it can be seen at

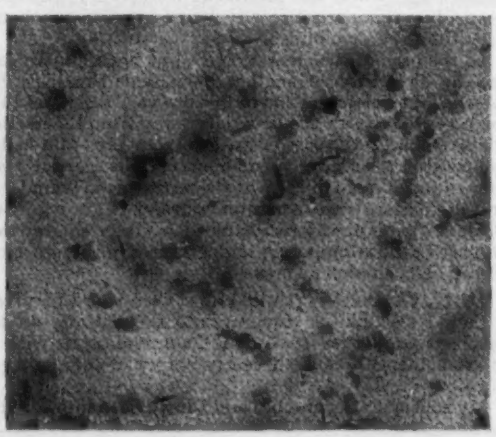


Fig. L—Tempered ¼ h at 700° C. Light initial etch × 200,000

once that there are apparently far fewer carbides, while those that are there appear to be in fairly regular orientation. There are one or two misfits, but this is inevitable; mostly they are correct.

Both these pictures were produced from a specimen tempered for 250 h at 700° C. We obtained the flat, plate-like carbides which Mr. Pickering has described, and this made us favour the plate rather than the needle school of thought. We then applied this technique to earlier stages of tempering, and Fig. *L* shows the structure obtained from a specimen which had had only ¼ h tempering at 700° C. The magnification of × 200,000 shows very convincingly that even at this early stage of precipitation there is a regular Widmanstätten pattern and that the carbides are coming down as fairly sharp-cornered square plates. In view of the suggestion in one of the papers about the ellipsoid shape of the precipitate in its initial stages, we should very much appreciate the authors' comments on these observations.

Dr. **D. McLean** (National Physical Laboratory): Anyone who has looked at high-power electron micrographs of alloy steels cannot fail to be impressed by the extraordinary variety of shapes of particles which are seen. A question which presents itself is whether the shape of the particle has any effect on the properties of the steel. I think that it does. Given a certain total volume of particles, if these are in the form of rods or plates then a dislocation will intercept more than if they have a more equiaxed shape. I think that this is fairly clear intuitively, and arithmetical analysis supports this conclusion. Disc and rod-shaped particles therefore stop slip more efficiently than do particles of other shapes. If the particles are surrounded by a region of strained matrix, rods in particular are still more advantageous, because the volume of strained matrix to the volume of particle is bigger for a rod than for any other shape of particle; and so, considering particle shape alone—and, of course, other factors are also important—a structure to aim at is a matrix containing coherent rod-shaped particles.

Dr. **B. A. Bilby** (Sheffield University): Professor Honeycombe has shown that additions of tantalum and niobium have a pronounced retarding effect on the growth of the V_4C_3 particles, and has suggested that these additions may have some effect on the diffusion rate. Another possibility has been examined in some work by Hillert on the rates of graphitization of steel which are retarded by molybdenum additions, and I think that we should bear in mind that a similar effect may be important here. The idea is that if the particle which is trying to grow is causing volume expansion, one way in which this expansion can occur is by deformation of the matrix. If these tantalum or niobium additions are made, they presumably harden the matrix, because the creep resistance is greatly increased, so that it may be that the growth of the particles is partly controlled by the increased resistance to deformation of the surrounding matrix, which is caused by the formation of the very finely dispersed carbides, or, in the case where only 0·05%Ta was added, perhaps by the effect of tantalum in solid solution.

Mr. **A. P. T. Taylor-Gill** (Edgar Allen and Co., Ltd.): I am much concerned with the practical application of modern physical metallurgical thought to the heat-treatment of high-alloy tool steels. Although I apply such practical yardsticks as estimation of the proportion of carbides remaining out of solution after austenitizing exposures conducive to maximum solution at minimum grain growth, together with practical study of the transformation response in terms of structure, hardness, and size, I have always wanted some accurate means of measuring the degree of diffusion of alloys and alloy carbides in austenite which has actually taken place for any combination of time and temperature during austenitizing.

Table I

ELECTRON DIFFRACTION MEASUREMENTS

Selected-area Diffraction Patterns at 80 kV, Philips E.M., Carbon Extraction Replica of Steel tempered 1 h at 700° C: Lattice spacings, Å

Former Retained Austenite	Former Martensite	A.S.T.M. Standards	
		V_4C_3	Cr_7C_3
2·40	2·40	2·40	
	2·30		2·29
	2·13		2·10
2·06	2·07*	2·07	2·03
	1·92		1·88
	1·83		1·84
	1·77		1·80
	1·61		1·75
1·47	1·47 s	1·47	1·43
1·39	1·35		1·34
1·26	1·26	1·25	
	1·19	1·20	1·20
	1·12		1·16
	1·025	1·04	
	0·985		
	0·955	0·952	

* Broad and intense

Dr. Nutting has spoken of the possibilities of micro-analysis by electron diffraction. Could this method be applied to give the information I require, as this might prove very helpful in assessing and controlling the hardenability of high-alloy tool steels.

Dr. **E. Smith** (Department of Mines and Technical Surveys, Ottawa): In their X-ray analysis of carbides from steels containing Cr and V, Shaw and Quarrell* do not report the presence of vanadium carbide in a steel containing 0·21%C, 9·161%Cr, 0·93%V. If V_4C_3 were present as a very fine precipitate, their methods may not have been able to detect it. Alternatively, after the long tempering times used by Shaw and Quarrell (500–2000 h) at 700° C, the equilibrium may favour the existence of only chromium carbide.

The authors, who have used the results of Shaw and Quarrell as a basis for their investigation of the effects of vanadium on a 9%Cr–0·2%C steel, do not report the presence of vanadium carbide in the early stages of tempering, although increasing the amount of vanadium produces an increase in secondary hardening.

I have examined, by electron microscopy and electron diffraction, specimens of a vacuum-melted steel of composition 0·2%C, 9%Cr, 1%V. The specimens were quenched into oil from 1200° C. The quenched specimens contained about 30% of retained austenite.

Electron diffraction patterns obtained from carbon extraction replicas of a specimen tempered for 1 h at 700° C, showed the presence of vanadium carbide (*see* Table I). Selected-area electron diffraction at 80 kV showed that the former retained austenite areas contained chiefly V_4C_3 with very little Cr_7C_3. The carbides of the former martensite were Cr_7C_3 and V_4C_3. The Cr_7C_3 was predominant but the rings of the V_4C_3 were the less spotty, indicating that the V_4C_3 had the smaller particle size.

Microhardness measurement showed that the ferrite gave a greater degree of secondary hardening than the former retained austenite.

* S. W. K. SHAW and A. G. QUARRELL: *J.I.S.I.*, 1957, vol. 185, pp. 10–22.

Since vanadium carbide has been detected as a very fine precipitate in this steel, which is similar to the authors' steel *D* (0·22%C, 8·99%Cr, 0·97%V), it appears probable that the precipitation of V_4C_3 rather than Cr_7C_3, with vanadium replacing chromium, is responsible for the secondary hardening observed when vanadium is added to a 9%Cr steel.

AUTHORS' REPLIES

Professor **Honeycombe:** In 9%Cr steels containing up to 1%V, tempered in the secondary hardening range 500–550° C, we did not detect V_4C_3, a finding which Mr. Pickering has confirmed. However, we certainly do not exclude the possibility of V_4C_3 forming at higher temperatures. Our steels were water-quenched to produce an almost fully martensitic condition, while Dr. Smith's steel had 30% retained austenite in which V_4C_3 predominantly precipitated at 700° C. The fact that he also observed some V_4C_3 in the martensitic regions after tempering at this temperature is by no means conclusive proof that V_4C_3 is responsible for a secondary hardening effect in the range 500–550° C. In this temperature range, with high-chromium steels, chromium appears to diffuse more rapidly in ferrite than does vanadium, as the formation of Cr_7C_3 after as little as 1 hour at 500° C suggests. We think that V_4C_3 cannot form as rapidly, and that the vanadium is primarily in solid solution in the Cr_7C_3. However, if the chromium content of the steel is markedly lowered, then V_4C_3 formation does occur as Shaw and Quarrell and Nutting have shown.

In reply to Dr. Goldschmidt, we have not investigated very short tempering times (< 1 hour) at 700° C, and so cannot say whether Fe_3C forms first or not. I think that at this temperature, diffusion rates are becoming so rapid that alloy carbides will probably form directly from the ferrite. I agree that the early stages of precipitation should be studied but preferably at lower temperatures than 700° C, when pre-precipitation effects would be more pronounced. Single crystal work would certainly be more conclusive in this respect but the difficulties in ferritic steels are formidable. We hope soon to study precipitation in single crystals of Cr–Ni austenites, as these alloys are more amenable to single crystal growth.

We are of the opinion that chromium carbide formation and cementite disappearance are partly interrelated as there seems to be good evidence for the formation of Cr_7C_3 *in situ* at Fe_3C particles. Nevertheless the martensitic grain boundaries are preferred regions for the early formation of Cr_7C_3, possibly due to the retention of a thin austenite layer in these regions. We agree that the iron carbide must go back into solution to provide carbon for the alloy carbide, for this is thermodynamically more stable, but can only form at higher temperatures when the increased rates of diffusion accelerate the approach to equilibrium.

There are undoubtedly changes in lattice parameter during the formation of the alloy carbides as the earlier work of Shaw and Quarrell has shown. In some cases there was evidence in our work of the formation of intermediate structures, but this question needs much more thorough examination.

While I would agree with Dr. Goldschmidt that the carbide–carbide *in situ* transformation is fundamentally important, I do not think it is as significant as the matrix–carbide precipitation as far as the mechanical properties of the steels are concerned. These properties depend on the readiness with which plastic flow occurs in the matrix, and the primary barriers to dislocation movement are the particles of carbide forming from the matrix. The maximum effect is dependent on several factors; we have referred to coherency stresses while Dr. McLean has underlined the role of particle shape. Dr. Bilby has referred to the deformation of the matrix by the growing carbide particles, and it must be admitted that in this phenomenon an *in situ* transformation could give rise to an increase in volume which would influence the mechanical properties.

I agree with Dr. Goldschmidt that there is some looseness in the usage of carbide formulæ and that not enough attention has been paid to the question of lattice defects. It is to be hoped that the newer physical methods will soon give us more detailed information on the compositions of various carbides at an early stage of the precipitation process.

We expected niobium and tantalum carbides to precipitate as one solid solution with vanadium carbide, but in our paper we indicated that although some solid solution took place, a separate Ta or Nb-rich carbide phase was frequently detected. However, with very small additions, about 0·05%Ta, this phase was not detected.

There has been some discussion on whether molybdenum and vanadium carbides in their initial stage of precipitation are rods, plates, or ellipsoids. My own feeling is that rods form first and grow into plates. Mr. Haynes' evidence relates to a steel tempered 15 min at 700° C which I consider to be in a rather advance stage of ageing. Further evidence is needed, preferably by direct observation of thin foils which have been tempered in the range 500–600° C.

Dr. **J. Nutting:** With reference to Mr. Pickering's comments, a further morphological point in deciding whether one has ε-carbide or cementite is that by doing selected-area diffraction one comes to the conclusion that cementite, if it is formed initially, is usually at the grain boundaries of the ferrite, whilst ε-carbide formed initially is a Widmanstätten precipitate in the grains. But at a later stage in the tempering one can get Widmanstätten precipitation of cementite, thus the transformation from ε-carbide to cementite is a monotropic one; it takes place *in situ*.

I agree with Mr. Pickering that molybdenum can go into solution in V_4C_3. We have now examined a steel of higher molybdenum content than any used previously and found evidence that the lattice parameter of the V_4C_3 is changing. We think this is evidence for molybdenum going into solution with V_4C_3.

We agree with Dr. Goldschmidt that the V_4C_3 lattice is probably not perfect, and it has been shown that by using selected-area diffraction techniques on the V_4C_3 in the ferrite grains one can get very peculiar streaking on some of the diffraction patterns. We believe that this comes from a superlattice defect of the type that Dr. Goldschmidt postulates.

With regard to the shape of the particles, Mr. Haynes suggested that there were clearly defined plates, but that was a case of tempering at 700° C: we are considering a very much earlier stage, when the vanadium carbide forms at 500° C. From other work we believe that plates can form at 500° C, and we can get changes in the diffraction pattern which we can interpret as changes of shape of the particles. We still think, however, that the carbon replica method is not very good in these early stages, and we hope to be able to give a more definite answer on the actual shape of the particles from the results of examining the thin films of the steel directly in the electron microscope.

In answer to Mr. Taylor-Gill it may be possible, by using the techniques of micro-fluorescene analysis or electron-probe analysis, to determine local concentration gradients in austenite.

DISCUSSION ON THE EFFECT OF PHOSPHORUS ON HIGH-PURITY ALLOYS

This discussion was based on the paper by B. E. Hopkins and H. R. Tipler entitled "**The Effect of Phosphorus on the Tensile and Notch-impact Properties of High-purity Iron and Iron–Carbon Alloys,**" which appeared in the March, 1958, issue of the *Journal* (vol. 188, pp. 218–236). Mr. **B. E. Hopkins** (National Physical Laboratory) presented the paper.

Dr. **B. Cina** (Brown-Firth Research Laboratories): As the problem of temper embrittlement can sometimes be serious in low-alloy steels, and as phosphorus has been suspected of being the main culprit, or at least a strong accessory after the fact, any work claiming to throw light on this deserves careful study.

However, there are certain features in the present results which one cannot easily reconcile with the known phenomena in temper-embrittled steels. Thus, while in plain Fe–P alloys it seems that the intergranular fractures occurred in the ferrite grain boundaries, the brittle intergranular fractures observed in temper-embrittled low-alloy steels occur in the prior austenite grain boundaries. Would the authors confirm whether all the intergranular failures observed in their Fe–P alloys occurred along the ferrite grain boundaries?

Secondly, the range of temper embrittlement in steels is approximately 350–600° C, whereas the range for embrittlement in the plain Fe–P alloys was, for some of the alloys, 600–800° or 600–1000° C. Thus we have a situation in which a temper-embrittled steel would be cured of its embrittlement by treatment at 700° C, the temperature at which the maximum embrittlement was obtained in the Fe–P alloys. Conversely, an Fe–P alloy embrittled by cooling through the critical range of 600–800° C could be cured of its embrittlement by further slow cooling below 600° C, where temper embrittlement would be obtained in low-alloy steel.

In dealing with a possible mechanism for the isothermal embrittlement of the plain Fe–P alloy, the authors invoke the idea of greater solute segregation to grain boundaries within the critical range of temperature than at higher temperatures. But that cannot easily be reconciled with the results of the athermal treatments, where it was found that a greater extent of embrittlement was obtained after quenching than after slow cooling from a temperature above the critical range.

It was found that the addition of carbon tended to prevent intergranular failures. Now, the higher-phosphorus alloys in the Fe–P series would have been largely ferritic at 950° C, while the corresponding Fe–P–C alloys would have been largely austenitic. Can the effect of carbon in preventing intergranular failure be due to the different structures obtaining in the two series of alloys at 950° C? This would mean either that phosphorus segregates less readily to austenite than to ferrite grain boundaries, or that carbon prevents segregation of phosphorus to austenite grain boundaries.

In the Fe–P–C series of alloys, embrittlement was found to be less rapid at 600° than at 500° C. Was it also less rapid at 700° than at 500° C? If so, this would mean that carbon was preventing segregation of phosphorus to ferrite grain boundaries at 700° C, since plain Fe–P alloys embrittle rapidly at those temperatures. It would also mean that carbon was encouraging segregation of phosphorus to ferrite grain boundaries at 500° C. Such divergent roles for carbon require explanation.

Finally, one thing that clearly emerges from the authors' work is that the effect of an alloying element on the mechanical properties of iron may not necessarily be the same when added singly and when added in the presence of other alloying elements. It is hoped that the authors will continue their investigations, particularly on the latter lines.

Mr. **G. Burns** (Department of Engineering and Materials Research, The Admiralty): I found the brittle intergranular fractures which the authors have developed in the phosphorus alloys interesting because I had seen some years ago some work by S. H. Rees which was published in the *Journal* in 1934. He was determining the mechanical properties of a very low carbon, high-phosphorus iron with practically no manganese. As air-cooled from 1000° C, that particular steel broke in a tensile test at room temperature at a stress of just under 30 tons/in², which one might expect, with about 6% elongation. Water-quenched from 1000° C, it broke at an extremely low stress, rather less than 13 tons/in², with no measurable elongation. In fact, that particular test-piece broke in two places simultaneously. As I remember the fractures now, they were coarsely granular fractures looking very much like the type that the authors got in their Fe–P alloys, but in this case with some carbon present and at room temperature and not at a very low temperature.

The suggestion that the intergranular brittleness developed in these alloys is in some way akin to temper embrittlement, and the finding that this embrittlement occurs in so short a time, interested me, because some years ago a colleague and I did a few rather rough tests to find out how long it would take a temper-embrittled steel to be restored to its tough condition. This was a Ni–Cr steel oil-hardened and tempered at 625° C and slowly cooled to get full embrittlement. We then took notched test-pieces, reheated to 600° C in a salt bath for varying lengths of time, quenched, and tested. We found that the time required to restore the material to full toughness was just under one minute, including the period of heating in the salt bath; so going from the brittle to the tough condition in our experiments took almost exactly the same time as to go from the tough to the brittle condition in the authors' experiments.

The microstructures in the paper interested me. Some of them were very reminiscent of effects that I have seen in low-carbon steels which have been quenched between Ac_1 and Ac_3, as does happen in the core of case-hardened articles. I remember that in 1944 Mr. Schofield described structures of this sort in a number of steels, and in the discussion on that paper Mr. Comstock said that he had noticed similar structures in the past and always ascribed them to the presence of phosphorus in the materials; so we have again a link.

One thing that I want to ask the authors is whether they had any more direct evidence of this phosphorus segregation to grain boundaries. I gathered from Mr. Hopkins' introduction that they have not. But from a conversation I thought they had been tackling this by another technique, and I wondered whether they had had any luck with it.

Professor **R. W. K. Honeycombe** (Sheffield University): I agree that the surface energies of the boundaries are lowered by the migration to them of phosphorus. However, it appears that the enriched layer is of appreciable thickness, rather thicker than we might anticipate in terms of equilibrium segregation. It is difficult to see why, in the presence of such high phosphorus concentrations, precipitation of a second phase should not occur. In other comparable cases, there is evidence of such a second phase resulting from the segregation of alloying elements.

We have examined cast Ni–Cr steels (e.g. 4%Ni, 1·4%Cr, 0·2%C) which show intergranular embrittlement particularly on annealing the cast steel in the vicinity of 600° C. Mr. I. S. Brammar has taken carbon extraction replicas from the fracture surfaces which have revealed the presence of very thin plate-like precipitates of manganese sulphide. These precipitates may vary in

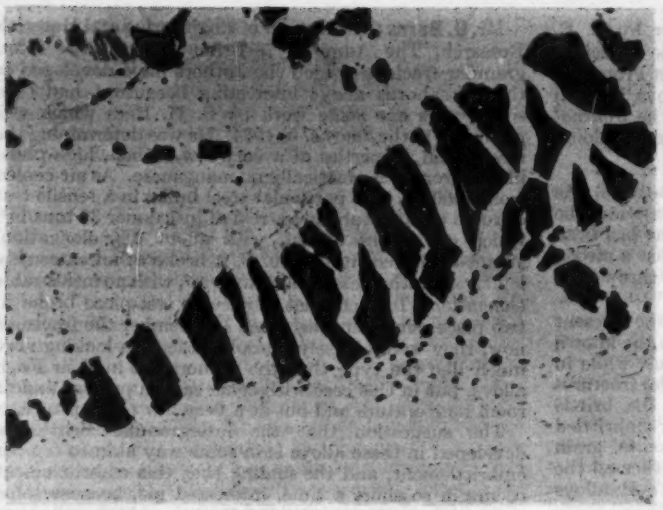
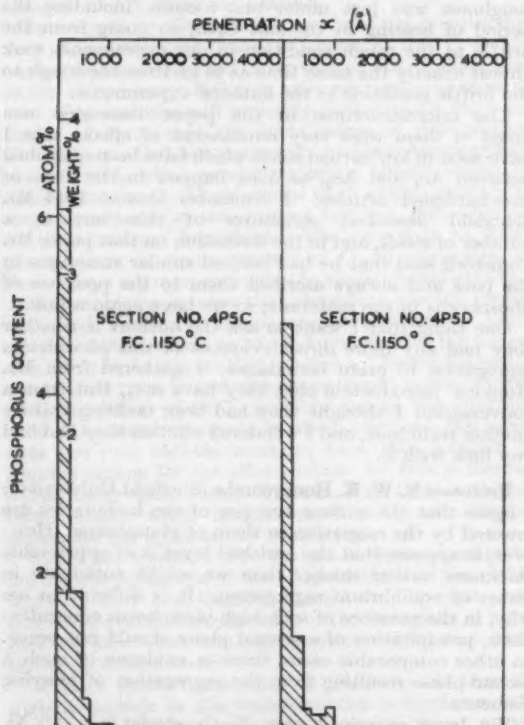


Fig. A—Carbon replica from intergranular fracture surface of a cast Ni–Cr steel × 12,000

Fig. C—Profile view of dome structures × 5600

length from 1 to 200 μ; a typical example is shown in Fig. *A*, but they are only about 1000 Å thick so they are difficult to detect in sections through grain boundaries. It seems quite likely that the fracture was initiated by these plate-like particles which were found to be very brittle.



F. C. = Furnace cooled

Fig. B—Histograms showing distribution of phosphorus at grain boundaries

I am not saying that similar precipitates occur in high purity Fe–P alloys, but it is a possibility which should be borne in mind. We have examined fractured specimens from the National Physical Laboratory with negative results so far, but there is the difficulty that a precipitate may dissolve preferentially in the etchant used to detach the replicas.

A second point which I would like to raise is that the brittleness in Fe–P alloys develops particularly on reheating air-cooled material to about 600° C. Why then did not the brittleness develop during the original air-cooling? One would assume that the material had spent sufficient time in the vicinity of 600° C for the embrittling reaction to take place. It may have disappeared in passing through the lower temperature range, but it is difficult to see why equilibrium segregation of phosphorus should fluctuate in the range 600° C to room temperature.

Dr. **M. C. Inman** (National Physical Laboratory): In reply to Mr. Burns, some typical results obtained by the radioactive tracer technique for the detection of grain-boundary segregation of phosphorus in the alloys under discussion are shown in Fig. *B*, which illustrates the high phosphorus content of two 'mirror image' grain-boundary fracture surfaces. The excess phosphorus, which appears to be almost wholly concentrated in a grain-boundary surface layer about 200 Å thick, is equivalent to between 2 and 20 monolayers of the pure element at the grain-boundary surface. Since an 'equilibrium segregation' should not exceed a maximum of about one monolayer, it seems likely that a large part of the observed segregation is due to the presence of phosphorus-rich precipitates at the grain boundaries.

An examination of the grain-boundary fracture surfaces with the electron microscope, using a carbon extraction replica technique, shows the presence of a large number of 'dome' structures on these surfaces. A picture of the domes as seen in profile is reproduced in Fig. *C*. Although no precipitate has yet been extracted inside these structures, we believe them to be the outline shape of a precipitate because they appear in opposition orientation (up and down) on the same replica.

Professor Honeycombe raised the question of recovery

of the alloys when quenched from temperatures below about 500° C. We could find no evidence of a diminution of phosphorus segregation at the grain boundaries as the quench temperature was reduced from 700° to 400° C, which is the behaviour one would expect if the phenomenon was purely 'equilibrium segregation.' However, in view of the possibility that a phosphorus-bearing precipitate is also present, the recovery of mechanical properties may be connected with a change in the form of such a precipitate or possibly with the growth of the precipitate at the expense of a segregated monolayer of phosphorus as the quench temperature is reduced.

Mr. **J. H. Gittus** (Mond Nickel Co. Ltd.): I have made some studies of temper embrittlement and the so-called recovery phenomenon in fully ferritic s.g. cast iron. The aim was to lower the impact transition temperature of the material without decreasing its hardness. The material was in one of two forms: it had either been slowly cooled during a period of 24 h from 700° C, which is the annealing temperature, to room temperature, or it had been water-quenched from the annealing temperature to room temperature. Equal numbers of specimens in these two initial conditions were subjected to isothermal heat-treatments at temperatures between 250° and 650° C, followed in every case by water quenching.

In the water-quenched specimens, embrittlement started earliest and proceeded most rapidly at 450° C. However, at 350° C the onset of embrittlement was preceded by a period in which toughening occurred. Further heating at this temperature of 350° C caused embrittlement, the process merely being delayed by the prior toughening reaction. In the specimens which were slow-cooled, the toughening reaction was absent and embrittlement proceeded throughout the temperature range under investigation, once more being most pronounced at 450° C.

The explanation of this phenomenon is that there are two processes operating during the cooling of this material: temper embrittlement and temper toughening. A specimen which has been slow-cooled from 700° C before being isothermally treated has passed through both temperature ranges. As it passed through the upper temperature range in the vicinity of 450° C it was embrittled. Subsequently, and at a lower temperature, it was toughened. The toughening reaction is not nearly as marked as the embrittlement, suggesting that in the slow-cooled specimens sufficient time was available in the vicinity of 350° C for the toughening to go to completion so that subsequent isothermal treatment in the toughening temperature range produced no further toughening. Accordingly, the process of embrittlement proceeded unchecked since this reaction, being more protracted and pronounced, had not achieved completion during the slow-cooling operation.

On this hypothesis, unless the materials are already fundamentally embrittled before the heat-treatment cycles are started, it would appear that there are two separate reactions which do not necessarily involve the same mechanism.

Dr. **D. McLean** (National Physical Laboratory): I should like to comment on Professor Honeycombe's remarks about precipitation in the grain boundaries. If a precipitate forms in equilibrium with a segregation in the boundary which itself is in equilibrium with the grain interior, then the precipitation is also in equilibrium with the interior. Therefore, precipitation of an equilibrium phase cannot occur until the grain itself is saturated. According to the literature, in the Fe–P system this requires 1%P at 600° C, whereas the alloys under discussion contained only about one-tenth this amount.

Mr. **A. M. Sage** (Union Carbide Ltd.): The authors compare the results obtained with the Fe–P alloys with those obtained earlier with Fe–N alloys. They may be interested in some results I obtained at B.I.S.R.A. on Fe–C–N–Mn alloys which may be compared with their Fe–C–P alloys. I found that increasing the nitrogen content from 0·002% to 0·016% in an alloy containing 0·17%C, 0·8%Mn air-cooled from 880° C raised the U.T.S. from 25 to 35 tons/in², the yield point from 15 to 25 tons/in², and reduced the elongation and reduction in area at 20° C; but at −192° C variations in nitrogen had little effect on the tensile properties. Thus the effects of phosphorous and nitrogen on the tensile properties appear to be similar.

We also found that the Charpy V-notch transition temperature was not significantly affected by nitrogen when the 20 ft-lb value was used as the transition-temperature criterion. When other criteria were adopted, such as 40 ft-lb, 50% crystallinity, 50% initial fibrous fracture, which are more comparable with the criteria which the authors have used, quite a large rise in transition temperature was observed. The addition of 0·010% N, for example, raised the 25% fibrous-fracture transition temperature 30° C. The reason for this rise in transition temperature when using these criteria was due to a progressive gradual broadening of the transition temperature range. Presumably the authors did not observe that broadening in the transition range with increase in phosphorus content. What are their comments on this? I would also like to know the type of fracture they observed in the fractured impact specimens.

My other point is prompted by Mr. Hopkins' introductory remarks. He suggested that the reason why one did not get embrittlement in isothermal treatment of the alloys containing carbon was that the carbon got to the grain boundaries first. I think I suggested this as a possible explanation for the influence of molybdenum in preventing temper brittleness in the discussion of a paper by Woodfine in 1954. At that time I suggested that segregation of the manganese and chromium atoms occurred at the austenitic grain boundaries and subsequently, during the temper-embrittling heat-treatment, atoms diffused to the ferrite grain boundaries, some of which were common with the austenite boundaries. I wonder whether this explanation would fit the apparent anomalies which have been referred to.

Mr. **G. M. Boyd** (Lloyds Register of Shipping): Some of the fractures have been of the intergranular and some of the cleavage type. Could the authors give some clue as to the causes of the discrimination between the two types of fracture, bearing in mind that most of the fractures that occur in ordinary commercial types of steels in the brittle condition are of the transgranular type?

I should also be grateful if they would give their definition of the transition temperature. I believe that they have stated this in some earlier papers, but I think it would be useful to have it in this context.

The other remark that I wish to make is a general one. This work on the alloying elements in pure iron arose from a suggestion by Sir Andrew McCance at a conference in Cambridge in 1945. Since then we have had a series of extremely interesting papers from the N.P.L. group, and on each occasion the question has been asked, either in private or in public: what connection has work of this sort with practical steelmaking and commercial steel? I think that that is a question that should not be asked of the authors. They are doing an excellent piece of controlled scientific work, but there is scope for some sort of liaison officer to interpret their results in terms of practical steelmaking.

It interested me very much to see that the carbon had a tendency to cancel the evil effects of phosphorus. As a

non-metallurgist, it has surprised me that if you put something into steel and it does harm, you can put something else in and cancel that harm. For instance, I am told that aluminium cancels the evil effects of nitrogen. I wonder whether the presence of many different substances tending to cancel each other could account for the behaviour of commercial steel being different from that of high-purity alloys.

AUTHORS' WRITTEN REPLY

The **Authors** wrote: Dr. Cina has remarked on some of the differences discussed in the paper between the characteristics of intergranular brittleness in Fe–P alloys and those of temper brittleness in commercial steels. In view of the differences in composition, it is not surprising that the range of temperature in which embrittlement occurs does not coincide in the two types of material. In particular, we should like to emphasize the pronounced influence of carbon in shifting the temperature range of embrittlement, and altering the rate and attainable degree of embrittlement of Fe–P alloys until they conform closely with those of temper brittleness.

It has for some time been accepted that with normal temper-embrittlement (lower-nose embrittlement as distinct from the upper-nose embrittlement of Libsch and co-workers,[1, 2, 3] and Jaffe and Buffum[4]), brittle fracture occurs predominantly along the prior austenite grain boundaries. In the Fe–P alloys, it is probable that a substantial proportion of the ferrite grain boundaries were previously austenite grain boundaries, particularly at the slower rates of cooling from the austenitic region. But the observation that the prior austenite grain boundaries are mainly concerned in the fracture of temper-embrittled steel does not necessarily imply that some change was brought about at these boundaries while the material was in the austenitic condition, as seems to be assumed by Mr. Sage in some of his remarks. Temper brittleness has usually been investigated in steels having a structure of tempered martensite, and for such a condition one would expect a general difference between prior austenite and newly formed ferrite grain boundaries arising from the mode of formation of the martensite. The orientation relationship between adjacent grains of ferrite separated by a newly formed grain boundary would be expected to make this boundary a less favourable region for segregation of solute atoms, or even precipitation, than the prior austenite grain boundaries separating regions bearing no crystallographic relation to each other.

There is no evidence in our work of a direct effect of lattice type on tendency to embrittlement such as would be shown up by isothermal treatments in the ferritic and austenitic regions, although there may be an indirect effect if substantial grain refinement occurs on going through the transformation $\alpha \rightarrow \gamma$ for treatment in the γ-range. Thus the 0·31%P alloy, for example, is seriously embrittled by isothermal treatment at 1000° and 1100° C where it is largely austenitic. It does not appear, therefore, that the influence of carbon on the embrittlement of the Fe–P alloys could be explained on these lines. The evidence indicates that embrittlement occurs in the Fe–P–C alloys at temperatures around 500° C in the ferritic range.

Most of the binary Fe–P alloys were not obtained in a condition completely free from embrittlement, the test-pieces having a comparatively high brittle fracture stress at — 196° C still breaking partly along the grain boundaries. The extent of this 'residual' embrittlement was probably not less than the maximum embrittlement obtainable in the Fe–P–C alloys by isothermal treatment at 500° C. Consequently we do not think there is any evidence that carbon encourages segregation of phosphorus at 500° C while inhibiting it at higher temperatures, as Dr. Cina suggests.

Both Professor Honeycombe and Dr. Cina refer to the recovery from embrittlement on cooling at different rates from 950° C. The effect of this type of treatment depends on the degree of embrittlement at 950° C, and the extent to which this is modified by cooling at different rates through the generally more severe embrittlement range at somewhat lower temperatures, and the recovery range at still lower temperatures. It does not appear that this is any more difficult to explain in terms of grain-boundary segregation than of precipitation.

The information given by Mr. Burns on embrittlement of low-C, high-P, low-Mn iron, and the very rapid recovery from embrittlement in a Ni–Cr steel by heating at a high tempering temperature is interesting. Similar results on recovery from temper brittleness have been given recently by Bhat and Libsch.[3]

Whilst we would agree with Dr. Inman that it is possible that the concentration of phosphorus at grain boundaries found experimentally might be at least partly accounted for by a precipitate, this has yet to be proved. Low grain-boundary strength is obtained in material freshly quenched from the embrittling temperature, so that it would be desirable to demonstrate the presence of a precipitate in this condition to avoid possible confusion due to ageing effects at room temperature. The traces of carbon present in the Fe–P alloys are sufficient to give ageing over a period at room temperature, and precipitation of carbide under these conditions has been demonstrated.

The observations of Mr. Gittus on temper brittleness in ferritic spheroidal graphite cast iron are of considerable interest because, although the embrittlement characteristics are so similar to those in steel that the phenomenon seems to be essentially the same in the two types of material, in cast iron there is the additional feature of recovery, which provides a link with the results on Fe–P alloys, as mentioned in the paper. However, we wonder whether the introduction of the term temper-toughening for recovery, or the reverse reaction to temper embrittlement, is desirable at this stage.

In reply to Mr. Sage, there was no marked broadening of the transition range in the Charpy impact test on increasing the phosphorus content up to 0·15%. Even with the 0·31%P alloy, the transition was largely spanned in 10–15° C. The effect of phosphorus on the transition temperature cannot, therefore, be accounted for in this way, as in the interesting results he presented for the effect of nitrogen in Fe–C–Mn alloys.

With regard to Mr. Boyd's remarks, phenomenologically, the choice of grain boundaries or cleavage planes for the course of a brittle fracture is governed by the relative values of the grain-boundary and cleavage strengths. From the metallurgical viewpoint, and without going into detail, it is considered that a reduction of the intergranular strength to below that of the cleavage strength could arise from the effect of solute atoms (or precipitates) on the surface energy. Finally, we should say that the question of interaction between various elements is a very large one. The investigation of some of these effects has been one of our objectives over a period of years.

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2. G. Bhat and J. F. Libsch: *Trans. A.I.M.E.*, 1955, vol. 203, pp. 330–335.
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DISCUSSION ON LOW-CARBON BAINITIC STEELS

This discussion was based on the following papers (the dates of publication in the *Journal* are given in parentheses):

"**Low-carbon Bainitic Steels,**" *by K. J. Irvine and F. B. Pickering* (1957, vol. 187, Dec., pp. 292–309).

"**The Metallography of Low-carbon Bainitic Steels,**" *by K. J. Irvine and F. B. Pickering* (1958, vol. 188, Feb., pp. 101–112).

Both authors took part in the presentation of the papers.

Dr. **N. P. Allen** (National Physical Laboratory): These two papers, reporting the planned development of a practicable bainitic steel, are of great importance, and also of great interest in demonstrating the vicissitudes through which a technological development may pass. When Bain first discovered the new constituent formed by isothermal transformation of plain carbon steels, he thought that this constituent was tougher for equal hardness than tempered martensite and tried to develop its applications accordingly. It was later shown that this was on the whole a mistaken view and that the best all-round properties in the medium-carbon heat-treated alloy steels were produced by tempering after a fully martensitic quench. The subsequent emphasis on 'hardenability' arose from this conclusion.

The inferiority of the tempered bainite structures was seen principally in the impact values at high hardness levels, but this was to a large extent a consequence of the carbon contents employed.

About the time that these matters were being discussed a grain-refined fairly high-carbon molybdenum steel was developed with about 0·6%C and 0·5%Mo. This was essentially a bainitic steel, and very good properties and excellent hardenability were claimed for it. Its properties in the fully hardened and tempered condition were indeed very good, but the impact value of the bainitic constituent formed when it was incompletely hardened was poor, and for some reason that steel did not appear to prosper.

About the time that that steel was being developed, I was interested in an investigation of the influence of the carbon content of steel upon the deterioration of impact properties resulting from transformation in the bainitic range, and found that this deterioration became progressively worse as the carbon content of the steel was raised, and was comparatively slight in low-carbon steels, so that it seemed desirable to be able to make bainitic steels of low carbon content; but the alloy content considered necessary to suppress the formation of primary ferrite appeared to be too high to be practicable, and it needed the discovery of the special effectiveness of a little boron in suppressing the primary ferrite to bring this type of steel to the fore again. It is very interesting to find after following that line of thought some years ago that the solution has come from another direction some years later.

Considering the applications which are in mind for this steel, and particularly welded structures and rotors, it is clear that the success achieved in maintaining the impact value of the steel will prove to be critical. It is unfortunate that in these papers so little is said about impact values. We look forward to the further paper which is promised on this subject. I gather, however, that the indications are that the Charpy V-notch transition temperatures of the steel are around 0° C and not very susceptible to modification. We are not told exactly how these impact transition temperatures have been defined, so that we do not know what interpretation to put on them. The question will arise of whether a Charpy V-notch impact transition temperature of about 0°C will be adequate for the sort of applications for which this steel is envisaged. It is becoming clear that the load-carrying capacity of a welded structure is very closely connected with its impact transition temperature, and consequently the answer to this question may well decide the ultimate usefulness of these interesting steels.

In the second paper, on the metallography of these steels, it is good to see such skilled use being made of electron micrographs, but when considering a case where so many features of the structure vary simultaneously it is very difficult to decide which of the features is the decisive one. I feel that there is a little over-emphasis of the effect of grain-size in the interpretation of the results, because the grain-size varies only between 1·5 and 2·5 μ over the whole strength range, and I can hardly think that this variation of grain-size can be very important. I am very much more interested, therefore, in the range of carbide particle size which the authors have observed, which is much wider. I fancy that the desirable properties of these steels are associated with the fine structure shown in Fig. 8*o* and *r*, where we have ferritic grains with the carbide particles precipitated uniformly through them. A steel entirely of that structure gives high strength and great toughness. What we have to do is to make that structure with a carbon content no greater that is necessary to produce the yield-point which we need. When we have done that we have made a practicable high-tensile tough steel.

Professor **R. W. K. Honeycombe** (Sheffield University): How do you define the grain-size? I find it hard to extrapolate from the electron micrographs to what I think you mean by grain-size.

Dr. **J. Nutting** (Cambridge University): I agree with Mr. Pickering on the importance of grain-size in controlling the properties at the higher transformation range, but at times I wonder why the grain-size is so important. In the paper which I presented earlier we thought that grain-size was important in the tempered steel, but there one could imagine that the grains had been nucleated and the grain boundaries had a fairly large orientation difference. In these materials, it seems from the way in which the ferrite forms, and from the micrographs, that there must be only a very low difference.

The authors refer to sub-grains, and by definition a sub-grain is something with a low orientation difference. How do these sub-grain boundaries hinder the movement of dislocations? The whole point about grain-size hardening is that it is the grain boundaries which control the process, and it is the distance over which the dislocations have to move before they meet the grain boundary which matters. In the case of a grain boundary with very low orientation difference, the strength required to force the dislocation through that, or to activate dislocation on the other side, is not so great as with a high orientation difference. From other photographs, as the grains form from the ferrite it is obvious that the impurities in the austenite are being pushed ahead and form an adsorbed grain-boundary film. Is it this which causes hold-up of the dislocations, which forms a barrier?

With regard to the impact properties of these materials, one might conclude that, if there is only a low orientation

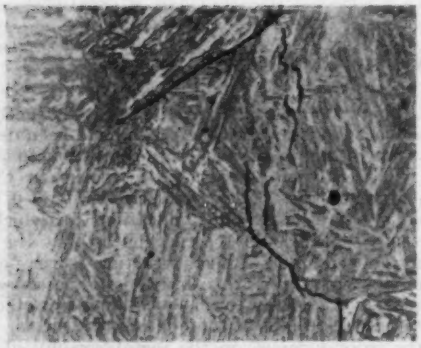

Fig. A—Cracking in the heat-affected zone of steel *B* (0·2%C–3·5%Ni–1·5%Cr–0·5%Mo); cooling rate at 300° C ≏ 10° C/s; max. hydrogen content of weld = 5·8 ml/100 g; bead-on-plate type test × 375

difference, once a crack starts to run through there is not much to stop it and if we want good properties we have to try to get something at the grain boundaries to stop the propagation of the crack through the material.

Mr. **J. H. Gittus** (Mond Nickel Co., Ltd.): Now that the discussion has turned to the question of impact properties, I should like to record some information which I have obtained with s.g. cast iron. We were concerned to produce fully ferritic s.g. cast irons having a very fine ferrite grain-size. It was already known that such refinement should produce a desirable reduction in the impact transition temperature. We made two approaches to this problem.

The procedure of ferritization involves first austenitization and then annealing at a temperature of, say, 700° C to graphitize the matrix. The size of the ferrite grains formed in this process is determined to a certain extent by the prior grain-size of the material, and in particular by the austenite grain-size. We find it possible to control the austenite grain-size by varying the austenitizing temperature. If this is 900° C, for example, then the ferrite formed by graphitization at 700° C will have an A.S.T.M. grain size of perhaps 5. If the austenitizing temperature is lowered to 800° C, then ferrite can be formed which has an A.S.T.M. grain size of 9.

Alternatively, ferrite can be produced by graphitizing bainite or martensite, instead of austenite or pearlite. Once again, a finer grain-size is obtained.

The finest ferrite grain-size was produced by the graphitization of bainite, and in certain experiments a grain-size as fine as A.S.T.M. 12 has been produced by this process. The grain-size produced by the graphitization of martensite was not as fine as that from bainite, but finer than that produced from pearlite. The graphitization of martensite, unfortunately, leads to the formation of additional tiny graphite spheroids which are situated at the ferrite grain boundaries and appear seriously to embrittle the matrix. The structures produced by graphitizing bainite, however, possess a fine, uniform ferrite grain-size, geometrically similar to that produced by the graphitization of pearlite. They are free from subsidiary graphite spheroids, the carbon from the matrix being precipitated on the pre-existing graphite spheroids. It is thus possible to reduce ferrite grain-size either by lowering the austenitizing temperature or by ferritizing bainite instead of pearlite. As expected, both methods produced an associated reduction in impact-transition temperature. However, in the former case, the maximum impact strength in the ductile range is unaffected whilst in the latter it is halved. No reason for this essential difference between ferrite formed from pearlite and that formed from bainite can be as yet ascribed. It strongly suggests, however, that the brittleness of tempered bainite in steel may be partly due to the structure of the ferrite constituent. Microscopic examination using a light microscope failed to reveal any particles in the ferrite matrix produced by graphitizing bainite which were not also visible in that produced by graphitizing pearlite. The only observable difference was the change in grain-size.

Nevertheless, the effect of this reduction in grain-size, which lowered the transition temperature in the way which we expected, has been accompanied by this unexpected reduction in maximum impact value, which was a great disappointment to us, because we hoped to use this heat-treatment on a commercial basis. It would therefore appear that there was some intrinsic feature of this ferrite, apart from its grain-size which gave it a low impact strength. It should be emphasized that in all cases the hardness of the fully graphitized material was virtually identical. Very large numbers of hardness determinations were made. On each side of the notch of every impact bar a small Brinell impression was made, to give an indication of hardness as close as possible to the actual fracture but not in the region of the deformed material and in no case was any significant difference in hardness shown.

Dr. **J. T. Berry** (Production Engineering Research Association): When I was associated with the Naval Construction Research Establishment, I became aware of the fact that although the mechanism of yield and fracture in mild steel was well understood, the same could not be said of that of flow and fracture in the low-carbon bainitic and low-carbon martensitic-type structures.

Dr. J. Heslop and Mr. E. Bagnall have concerned themselves with flow and proof-stress measurements in quenched and tempered low-carbon steels.* An interesting finding of theirs was that an abnormally high ratio (1 : 3) of flow stress to 0·1% proof stress existed for two of the as-quenched materials examined (a low-alloy steel and a high-manganese mild steel). Have the present authors carried out any measurements which would enable them to say whether a similar sort of ratio exists for any of the low-carbon bainitic structures?

Secondly, in the course of a study of certain fundamental aspects of weld metal and heat-affected zone cracking in low-carbon low-alloy steels,† which included an examination of the morphology of cracking and compilation of cracking diagrams relating the onset of cracking to hydrogen content and cooling rate at 300° C, a sensitivity to many microstructural details was noticed.

Figure *A* shows a typical heat-affected zone crack × 375. It can be seen that, although there is a considerable amount of inter-lath cracking (either between parallel laths or packs of laths), there exists some trans-lath cracking. Weld-metal cracking showed similar tendencies.

Mr. R. C. Allan and I think that, where the cracking has involved lower-bainitic laths, it has confined itself to interfaces of the ferrite–carbide type as they occur at the lath edge or within laths (i.e. 'Widmanstätten' carbide). No doubt a similar but more limited argument can be advanced for quench-tempered martensite.

The problem of fracture in such structures becomes similar to that of temper-embrittlement, i.e. a *cohesion* problem. Have the authors made any study of the mode of fracture of the steels whose structure and properties they have so adequately described?

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* J. Heslop and E. Bagnall: Unpublished work.

† J. T. Berry and R. Allan: Unpublished work.

Mr. **A. G. Haynes** (The Mond Nickel Company Ltd.): In considering the effect of boron on the isothermal transformation in the temperature range of pearlite formation, I was intrigued to see from Fig. 3 of the authors' December paper that the nose on the transformation diagram had apparently been raised approximately 100° C by the addition of boron to the ½%Mo steel (Fig. *B*).

In the boron-free steel, the nose occurs near the estimated Ac_1 temperature but, in the presence of boron, the nose occurs at a temperature approaching the Ac_3. The authors do not indicate in their isothermal transformation diagram curves showing the time required to form given percentages of ferritic and pearlitic transformation products, but it is obvious that the nose for the onset of carbide formation cannot be displaced to a similar temperature above Ac_1.

In studies of the isothermal transformation characteristics of numerous low-carbon alloy steels we have always found that the nose temperature of the pro-eutectoid ferrite formation curve is never markedly different from that of the nose associated with the onset of carbide formation in the later stages of the reaction.

The relative displacement of the noses for ferrite and carbide formations in boron-containing steels therefore suggests that the effectiveness of boron in retarding ferrite nucleation varies at different temperatures in the pearlite range and, unlike the effect of the common alloying elements in low-carbon steels, it does not promote a uniform displacement of the diagram to longer transformation times.

At high temperatures in the pearlite range where ferrite nucleation is mainly limited to a few sites at the junctions of three or more austenite grain boundaries, there is little difference between the times required to initiate ferrite formation in the boron-containing and the boron-free steels. At lower temperatures microscopic evidence shows that there is an increased tendency for ferrite nucleation to occur at junctions of three austenite grain boundaries and also between two neighbouring grains. At some temperature in this region boron appears to become effective in retarding nucleation, an effect which increases with further decreases of temperature, thus resulting in the appearance of a high-temperature nose. On the other hand, the boron-free steel still shows an increased nucleation rate on further lowering of the temperature to below Ac_1.

An earlier paper of the authors' indicated that the nose temperature decreased with decrease of boron content and this suggests that for each boron-containing steel there is a specific temperature above which there are insufficient interstital boron atoms effectively to reduce the grain boundary energy at the high-temperature nucleation sites. Is it also possible that progressive additions of molybdenum from 0 to ½% in a boron steel may alter the dispersion of boron at the austenite grain boundaries giving the rather unexpected increase from 620° to 790° C of the nose temperature for pro-eutectoid ferrite which was observed by Dr. Irvine and his colleagues?

I feel that if the authors would provide details of the progress of transformation in the pearlite range of the boron-containing and boron-free steels, such data would throw useful light on some of my comments.

I hope the authors will forgive my confining my remarks to the pearlite transformation range when we are discussing bainitic steels, but after all it is this reaction, or rather its absence, which permits attainment of relatively high tensile strength in the absence of conventional alloying elements.

In work on low-carbon, low- and medium-alloy steels of different hardenability, we have noted a similar relationship between the proof stress value and tensile strength to that given by the authors (Fig. 6 of the December paper). Tempering after normalizing or quenching effects a substantial improvement in the limit of proportionality and 0·1% proof stress values and to a lesser degree in the 0·5% proof stress. This applies in particular to those steels which do not wholly transform during cooling in the pearlitic range.

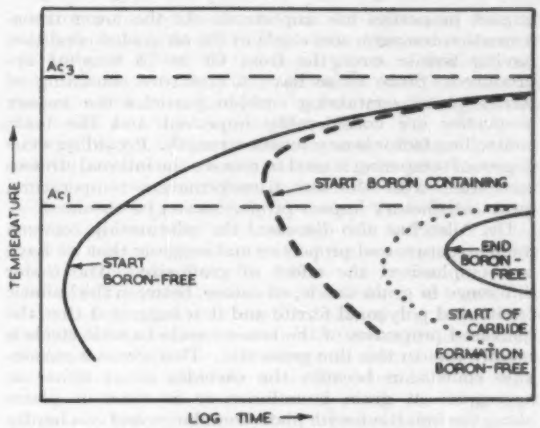


Fig. B—Transformation in the temperature range of pearlite formation

AUTHORS' WRITTEN REPLY

Dr. **Irvine** and Mr. **Pickering** wrote: Dr. Allen has raised the question of impact. Further work has been carried out on this subject and will be reported at a later date. The important points which emerge from this work can be given quite briefly, however. If the bainitic structures are examined it will be seen that they fall into two main types. At the higher transformation temperatures there are grains containing bainitic ferrite plates and the carbides occur either as large aggregates at the grain boundaries or as films or plates along the bainitic ferrite plate boundaries. At the lower transformation temperatures the structure consists of bainitic ferrite plates containing small precipitated carbide particles. In the high-temperature bainitic structure the easiest fracture path is the individual ferrite plate itself, and since this generally extends across the austenite grain, the austenite grain-size will be one of the controlling factors on impact properties. In the lower-temperature bainitic structures, since the ferrite plates contain carbide particles, the fracture path is very much interrupted and grain-size no longer has such a controlling influence.

With mild steel it is known that providing reasonable grain-size control is exerted, acceptable impact properties can be obtained. With the high-temperature bainitic structures, since the tensile strength has been increased quite considerably, it is necessary for the grain-size to be reduced in order to retain a similar level of impact properties. In general terms, if satisfactory properties can be obtained in a mild steel using a finishing rolling temperature of 1000° C then this must be decreased to 900° C for a 40 tons/in² bainitic steel and to 800° C for a 50 tons/in² steel. Because at intermediate transformation temperatures the tensile strength is still increasing (i.e. of the order of 55 tons/in²) but the structure has not changed to the type where carbide particles are within the ferrite plates, it will be expected that these steels will have poor impact properties even when grain-size control is exerted. Generally speaking therefore, these intermediate structures are not recommended where

impact properties are important. At the lower transformation temperatures steels in the air-cooled condition having tensile strengths from 60 to 75 tons/in² are produced. Since these have a structure consisting of ferrite plates containing carbide particles the impact properties are considerably improved and the main controlling factor is now tensile strength. Providing some degree of tempering is used to remove the internal stresses associated with the low transformation temperature, quite satisfactory impact properties can be obtained.

Dr. Allen has also discussed the relationship between microstructure and properties and suggests that we have over-emphasized the effect of grain-size. The major difference in grain-size is, of course, between the bainitic ferrite and polygonal ferrite and it is suggested that the increased properties of the lower-tensile bainitic steels is mainly due to this fine grain-size. This seems a reasonable conclusion because the carbides occur either as aggregates at grain boundaries or as films or plates along the bainitic ferrite plate boundaries and can hardly be expected to have much greater contribution to the strength of the steel than the grain boundary itself. Through the bainitic transformation range it is true that the grain-size variation is small (although nevertheless important), but it is at this stage that the carbide particles begin to be precipitated within the ferrite plates when they must result in dispersion hardening.

Professor Honeycombe has asked how the grain-size of these structures is measured. The grain-size is defined by the white grain boundaries of prints taken from photographs of Formvar replicas. These include not only the prior austenite grain boundaries but also the bainitic ferrite boundaries. The method employed is an intercept measurement and it is accepted that this measurment is difficult. For one thing, transformation occurs over a range of temperature of approximately 100° C and therefore there will be quite a variation in the grain-size. Another point is that the ferrite plates are elongated which always makes grain-size measurement difficult.

Dr. Nutting has also discussed the reason for the strength of these steels. We agree with him that it is difficult to envisage more than a slight orientation difference between the ferrite sub-grains. However, the energy necessary to force dislocations through these sub-grain boundaries will increase as the grain-size decreases. We also consider that the segregation of impurities to these sub-grain boundaries will create a strong barrier to the movement of dislocations. Consequently, the strength of the steel results from a combination of these two effects. We have already pointed out that at the lower transformation temperatures the dispersion hardening effect of the precipitated carbide particles plays an important part.

The results described by Mr. Gittus, although interesting, are difficult to relate to our work although they do appear to indicate the effect of ferrite grain-size.

Dr. Berry asked whether we have made any measurements of the ratio of flow stress to proof stress values. We are not sure how Dr. Berry defines flow stress but if he means the limit of proportionality then the measurement of this depends to a great extent on the sensitivity of the apparatus which is used and consequently, the ratio of flow stress to proof stress which is obtained will be quite variable. We have not attempted to make any measurements of this type. Dr. Berry also asked whether we had noticed any temper embrittlement effects in these steels. In our fracture studies we have noticed a tendency to fracture along the bainitic ferrite boundaries, particularly when the structures form at intermediate transformation temperatures when there is a tendency for carbide films or plates to occur at these boundaries. This type of fracture is restricted to these particular structures, however, and is not a general feature of the fracture of these steels. It will be remembered that these are molybdenum and boron steels, and although it is sometimes suggested that boron promotes temper embrittlement it is also known that molybdenum is effective in preventing temper embrittlement.

Mr. Haynes has made some comments on the high-temperature region of the isothermal transformation diagrams. We agree with his comments that in general, boron-free steels show the nose of the ferrite and pearlite C-curves at about the same temperature and that since in a boron steel the nose for the pearlite curve is not raised above Ac_1 there is a difference between the effect of boron on the ferrite and pearlite transformations. The reason for this is that boron does not uniformly retard the pro-eutectoid ferrite reaction over the complete temperature range. In fact, the segregation or adsorption of boron to the austenite grain boundaries is temperature dependent. At high temperatures near to the Ac_3 temperature the adsorption is small and, consequently, the pro-eutectoid ferrite reaction is only slightly retarded. As the temperature falls the adsorption becomes more pronounced and the pro-eutectoid ferrite reaction becomes markedly retarded. This results in the nose occurring at a temperature between Ac_1 and Ac_3. This means in fact, that the steel alters radically with regard to its boron dispersion as the isothermal transformation temperature decreases below Ac_3.

DISCUSSION ON TRANSFORMATION CHARACTERISTICS

This discussion was based on the paper by W. Steven and D. R. Thorneycroft entitled "**Variations of Transformation Characteristics within Samples of an Alloy Steel,**" which appeared in the September, 1957, issue of the *Journal* (vol. 187, pp. 15–32). Mr. **D. R. Thorneycroft** (Mond Nickel Co., Ltd.) presented the paper.

Mr. **P. W. Bygate** (Brown Bayley Steels Ltd.): I should like to congratulate Dr. Steven and Mr. Thorneycroft not only on the work they have put in, but also because they have attempted to measure systematically some of the differences in the behaviour exhibited by specimens from different portions of the same mass. As far as I know, the reason for the hardenability variations is not known and this is the first attempt to throw some light on the problem.

With regard to the 9 s.w.g. wire specimens isothermally transformed at 550° C for 24 h, it is said that the hardenability pattern, as shown in Fig. 2, was visible on a macroscale, but not on a microscale. It is difficult to understand why there is no difference microscopically between the outside and centre zones in view of the subsequent isothermal transformation data presented, unless it is due to a bad choice of transformation temperature, i.e. one at which over a very long period of time there is only a small amount of transformation, and this would probably be the case, according to the isothermal diagram shown in Fig. 5. Had these specimens been transformed at either the pearlite temperature of 640° C or the bainite temperature of 400° C—the temperatures used subsequently on the 2-in. square bar and ingot material—differences on a microscale might well have been much more readily apparent. It would also be of

interest to know to what extent, if any, the authors consider the transformation data, obtained dilatometrically for the specimens machined from wire, might be influenced by the fact that the specimens, judging from Fig. 2, would be inhomogeneous and to varying degrees.

Fig. A—Sulphur print of a section through the axis of a 25-cwt ingot of En 16 steel

It is noted from Table IV that the minimum total reaction time for material from the columnar zone of ingot 5 is 14 h at 610° C, yet for the same material transformed at 640° C the times given in Table V are only 1·4 and 4·2 h. The agreement for the material from the columnar zone at the bottom of the ingot in Tables IV and V is very much better.

With regard to the form of the zone of lower hardenability, we have on many occasions examined sections of billets or bars corresponding with the top, middle, and bottom ingot positions and, without exception, have found the variation recorded by the authors, i.e. the zone is large at the bottom and becomes progressively smaller towards the top. That being so, the low-hardenability zone can be visualized as an elongated or truncated pyramid within the body of the ingot. We also confirm that this variation in hardenability cannot be accounted for by variations in composition on a macroscale. The feature normally present in an ingot which corresponds approximately with the suggested pyramid outline is the zone delineated by the 'A' segregates. Ingots of the size and type examined do not generally show marked 'A' segregates, but sulphur prints may show an indication of some structural irregularity where the 'A' segregates would normally be expected. The sulphur print shown in Fig. *A* is from an ingot of the same size and weight as those used by the authors and the 'irregularity' in sulphur prints is plainly visible. It will be seen that this line of 'irregularity' marks the limit of the columnar crystals only at the bottom of the ingot (Fig. *B*) and that higher up, towards the top of the ingot, between the end of the columnar crystals and the line of irregularity there is a zone of equiaxed dendritic crystals which appear different from the crystals within the pyramid. Actually these structural changes are visible on the original print from which Fig. 13 was made, but the structural detail has been lost in reproduction, although indications of the zone are still visible, particularly about the mid-portion of the ingot. The tapered zone shown on the sulphur print corresponds with the tapered zone of the modified ingot structure referred to by the authors and stated to correspond closely with the limits of the zone of low hardenability.

This variation in ingot structure is not new. It was noted on some of the ingots examined by the Heterogeneity Committee, about one of which—a Ni–Cr–V steel ingot—Dr. Hatfield* has written: "A characteristic of this ingot which is confirmed is the complete absence of 'A' and 'V' segregates, although it might at first sight appear from the macro-etched surface that there was an 'A' segregate. Closer examination, however, reveals that this is in reality only an etching effect produced by a change in the character of the crystallization in this region." Subsequent comment states that a secondary system of equiaxed crystals has developed between the columnar zone and the inner zone of fine equiaxed crystals."

There appears to be insufficient evidence to justify the contention that the low hardenability zone is invariably associated with—to quote Dr. Hatfield—the inner zone of fine equiaxed crystals, but there is certainly a remarkable coincidence of these two features, for which reason I strongly support the authors' suggestion that these effects arise in the solidification of the ingot and that the accumulation of detailed knowledge regarding this solidification is of prime importance. In this connection the authors suggest that a study should be made of such variables as pouring speed, ingot size, etc. to ascertain their effect on the resulting ingot structures. We have

* *Iron Steel Inst. Spec. Rep.* No. 25, 1939.

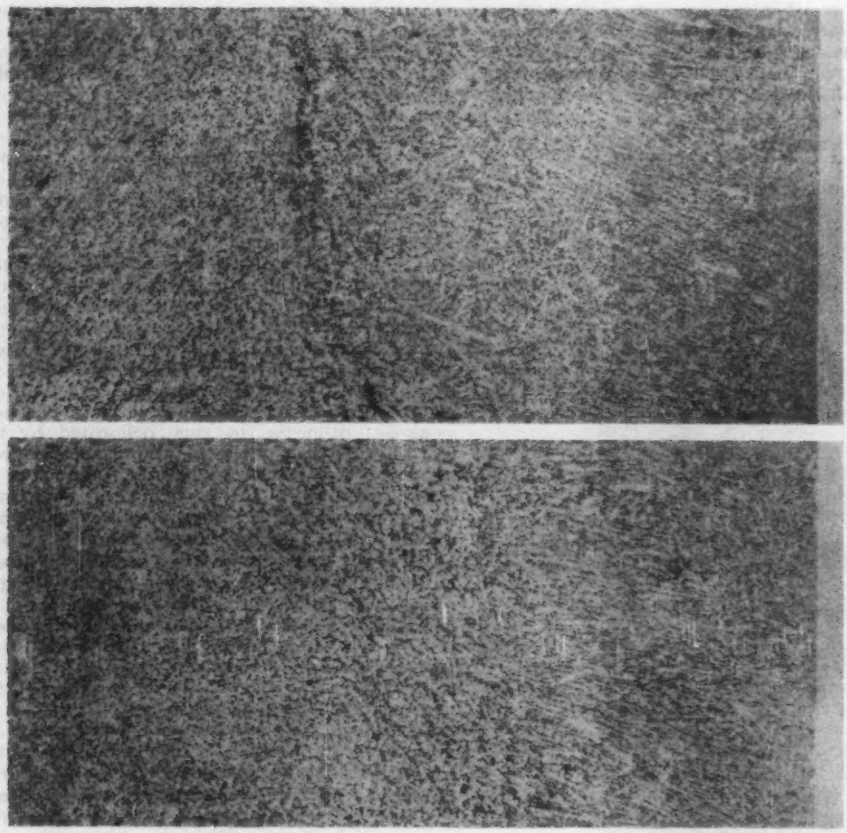


Fig. B—Portions of the sulphur print shown in Fig. A at positions towards the top and bottom of the ingot

carried out experiments involving several of the variables mentioned by the authors and as a result we feel that although the shape of the zones of varying crystallization can be modified, their suppression cannot be effected by any of these factors, that is, in ingots of the size now under consideration.

It seems, therefore, that there may be equiaxed crystal zones of varying hardenability, but those of low hardenability are selectively situated within a restricted portion of the ingot body. The two main zones may or may not be separated by a zone of peak hardenability, depending on the degree of 'A' segregation. If this is correct then it would appear important to determine how the two systems of equiaxed crystals arise, and this brings us back to the belief that the solution of this problem lies in the study of the basic mechanism of solidification of the ingot.

One can imagine how these zones may be built up embodying the occurrence of falling crystallites as postulated by the authors, but it is debatable whether the addition of inoculants is likely to prove a useful line of attack, since there appears to be no evidence to suggest that this inner zone of equiaxed or nuclear crystals (as they have been called) owes its occurrence to lack of crystallization centres. It is possible, however, that inoculants may cause an earlier onset of crystallization which may well affect the freezing pattern.

The effect of the homogenization treatment is very marked and the authors conclude that this is due to diffusion producing almost complete chemical homogeneity. They state that the microstructures of homogenized material are free from banding. If that is so it would be of interest if the authors would comment on the persistence of the fluctuating hardness giving a saw-tooth effect for material that has been homogenized and subsequently isothermally transformed. The fluctuating hardness must surely be a reflection of microcomposition differences and it is therefore only in respect of the removal of the centre dip in the hardness traverse curves that the homogenized material differs largely from the original. I venture to suggest that this saw-tooth effect is an indication that the microhomogeneity is not quite as complete as the authors conclude.

In terms of mechanical properties, the commercial importance of these hardenability variations is perhaps not very great since most engineering steels are tempered at such temperatures as largely iron out the as-hardened differences, but on occasion the low hardenability zone appears to reveal itself in other ways. Figure *C* shows the macrostructure along the centre line of a 4½-in. dia. bar of En 30 steel. Structural differences are visible which seem to portray the zones of different hardenability and it is evident that the inner, low-hardenability portion is subject to cracking and that the major cracks are clearly limited by it. Do the authors consider that the varying transformation characteristics they have ob-

served could account for this effect? The close correspondence of transverse cracks with a zone of low hardenability has been observed in several instances.

Mr. **J. H. Gittus** (Mond Nickel Co. Ltd.): It has been suggested that some type of inoculation might possibly influence the degree of microsegregation which occurs during solidification. This is an attractive proposition. Evidently the process of solidification is one in which segregation occurs, and diffusion to a limited extent will tend to reduce that segregation during the process of solidification and during the subsequent cooling of an ingot or a steel casting to room temperature.

The authors have demonstrated that homogenization may be achieved by protracted high-temperature heat-treatments. It was thought that if the rate of nucleation of the dendrites during solidification could be substantially increased, then the distance over which diffusion must occur to remove the microsegregation would be correspondingly decreased, since in the simple case this diffusion distance is equal to half the thickness of the crystallite. One way of increasing the rate of nucleation is to increase the rate of solidification, but unfortunately this also decreases the time available for diffusion at any specific temperature. What is therefore wanted is a means of nucleating a melt which does not rely for its effect on change in solidification rate.

In the absence of any knowledge of a specific inoculant for an En 24 steel, it was decided to use vibration. Because of practical limits involved in the application of vibration, it was impossible to study ingots of any size and use was made instead of 2½-in. dia. balls cast in graphite moulds. A consequence of the rapid solidification rate of castings of this description is the virtual absence of any equiaxed zone from the casting and the presence instead of columnar austenite dendrites throughout. The dendrites are orientated more or less perpendicularly to the surface of the casting in the vicinity of the cast surface, and this orientation is gradually lost as the centre of the casting is approached.

Two castings were made simultaneously in En 24 steel. One ball was cast into a vibrating mould and the other into a stationary one. The castings were then examined at the centre position remote from the small amount of centre porosity which was encountered, and various etching techniques at low magnification were used to demonstrate visually what we might term the macromorphology of the crystallites and, secondly the microsegregation. It was found that the vibration had reduced the size of the dendrites, eliminating their columnar appearance. Microsegregation and 'coring' of the dendrites was substantially reduced.

The **Chairman** (Dr. H. H. Burton): What was the frequency of vibration, and how was it applied?

Mr. **Gittus**: The frequency was 100 cycles a second, and the whole mould and the casting which it contained was vibrated up and down in one plane.

The **Chairman**: Various experiments of vibrating have been carried out on ingots, but so far I do not know of a single case where there has been a beneficial effect. I am not speaking of supersonic vibration experiments.

All the ones that I know about were carried out by Dr. Northcote at Woolwich, and they have had no beneficial effect. Some of you may remember a paper in the Institute given by my friend W. J. Dawson, who solidified an ingot on a jar-ram machine in the foundry and got a very marked segregate pattern which would not have occurred in the same sized ingot without vibration. We have carried out experiments on a series of ingots, all bottom poured, and we got entirely negative

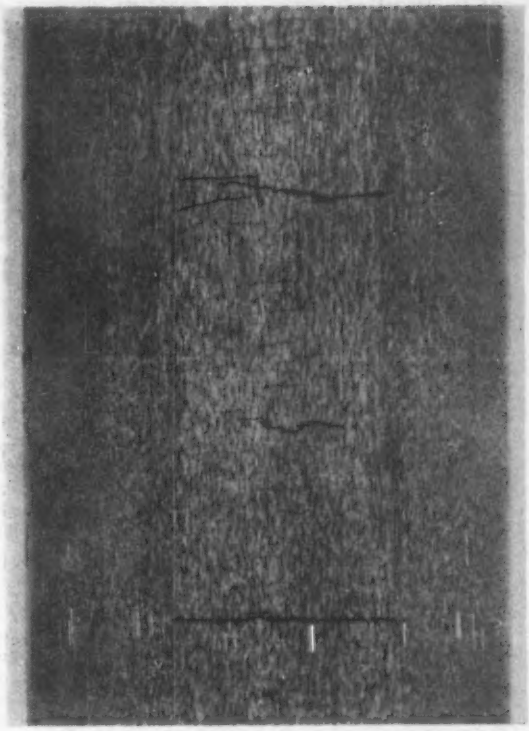
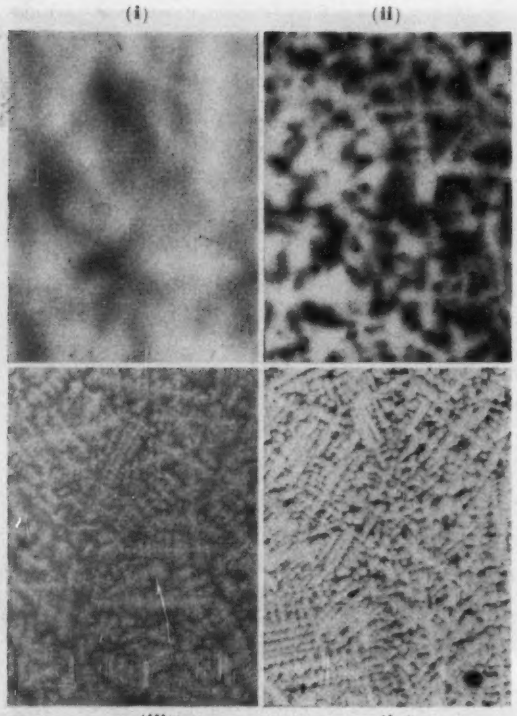


Fig. C—Etched section through the centre line of a 4½-in. dia. bar of En 30 steel, showing the confinement of transverse cracks to the low-hardenability zone

results. So I think that probably the frequency and the type of vibration may be very important.

Mr. **Gittus**: That may be so. I have no evidence about frequency, since it was not a variable in my experiments, but there is a paper before the Institute, proposed for publication, which describes this experiment and a number of others which throw some light on the matter of segregation which may have been caused by vibration.

Monsieur **A. Kohn** (IRSID, France): As the question of microsegregation has been raised, I can give some information on that important problem. At IRSID, we are investigating the various phenomena which take place during solidification of alloys, in order to achieve a better understanding of the mechanism by which segregations are produced. In the first part of that study we used the autoradiographic method to see in which conditions certain elements added to iron may segregate at the microscale. This study was carried out on elements which can be made easily radioactive by irradiation into an atomic pile: they include some elements which can be found in conventional structural steels, and others which are not usually present in normal commercial grades. Two series of alloys were prepared by remelting *in vacuo*, or under an argon atmosphere, pure iron with a master alloy of the desired element. The first series consisted of binary alloys, each comprising one of the following elements added in small amounts to iron: 0·025%P, 0·15%Cu, W, and As; 0·5% Mn and Co, and 2%Mo. The second series consisted of ternary alloys, in which the second element was as indicated above, and where the third element was carbon



(i) Segregation of phosphorus in a 0·030%P alloy without carbon
(ii) Segregation of phosphorus in a 0·015%P alloy containing 0·46% C
(iii) Segregation of copper in a 0·30%C–0·20%Cu alloy
(iv) Segregation of chromium in a 0·50%C–0·40%Cr alloy

Fig. D—Photomicrographs showing segregation of phosphorus, copper, and chromium × 5

at a content of about 0·40%. Ternary alloys of chromium (0·5%) and gold (0·01%) were also investigated.

The ingots, weighing about 1200 g, were cooled in the refractory crucible where they had been melted, in such conditions that the rate of cooling solidification was approximately similar to the rate which occurs in the inner parts of large industrial ingots.

Autoradiographs were made on samples about 1 cm × 1 cm, which had been thinned by polishing down to about 100 μ. The examination of the autoradiographic images has shown that a slight segregation of phosphorus and arsenic could occur in the alloys without carbon; but in the other binary alloys no segregation of the solute element could be detected. On the other hand, in each alloy with carbon—except in the alloy containing cobalt—a very distinct pattern, corresponding to a dendritic segregation, was observed: this heterogeneity was very clearly seen in the alloys with phosphorus or arsenic, thus confirming previous investigations. From these results (Fig. *D i–iv*) it can be deduced that most of the elements may segregate when they are present in an iron alloy containing carbon. Further tests are planned to confirm whether cobalt does not actually segregate. As the resolution power of the autoradiographic images is poor, it cannot be concluded that no segregation was present in the alloys without carbon, where it was not observed; but, if such a segregation exists, it is very slight.

The effect of the carbon on the intensity of the segregation of an element is probably related to the broadening of the solidus–liquidus range, produced by the carbon.

Thus, the tests give direct evidence that probably all elements which are present in structural steels as alloys or as impurities may segregate. Up to that time, qualitative information only has been drawn from examination of the autoradiographs; therefore, by comparing the images, it can be deduced that arsenic, phosphorus, and gold segregate the most. But, at IRSID too, Mr. Philibert, using the electron probe spectrometer, was able to detect high variations in the content of some elements in commercial air-hardening steels. The variation was about 35% for copper, 50% for nickel and chromium, 65% for manganese, and 130% for molybdenum. Then it is not surprising that a high variation in the tempering response may be observed, when examined on a microscale.

Mr. **E. Johnson** (Wm. Jessop and Sons Ltd.): Some three years ago we investigated, in a similar manner to that of the authors, the variations of transformation characteristics of steel billets of En 17 as well as En 24 quality; and found large transformation variations in both longitudinal and transverse directions of both steels. Transformation in the pearlitic zone showed the effect more markedly, and comparison of the bainitic and pearlitic characteristics of various portions of the billets revealed no uniformity on either side of the longitudinal axis, nor a gradual progressive change from top to bottom. In general there is a good correlation between the two results on En 24, and like the authors we were not able to account for the intermediate zone of peak hardenability on the basis of results of macro- and micro-examination and spot spectrographic determinations. Further work is therefore indicated.

With regard to transverse cracking, the only comment I have to make is that the phenomenon is definitely related to hydrogen content. In a works experiment we took half the billets of one cast and by slow cooling (and presumably getting rid of the hydrogen) avoided the effect in the subsequent bars.

Mr. **F. G. Haynes** (Mond Nickel Co., Ltd.): I should like to suggest one direction in which there may be some commercial importance in this phenomenon of microsegregation. I want to talk about a very simplified case which does not represent the practical case very closely, but it will indicate the point I wish to make.

Supposing that a dilatometer specimen has a laminated structure composed of alternate layers of 'A' material and 'B' material and that the 'A' material is less rich in alloying element, so that it starts transformation first. On cooling down, or indeed on holding at constant temperature, the 'A' material, transforming first, will try to expand. It will be very weak because a metal undergoing a phase change has very little strength. Therefore, being trapped in between the layers of 'B' material which are not yet transforming, it may well get upset a little. Thus, when it has finished transforming, it will be a little shorter than it should be. Similarly, when the 'B' material starts to transform, a shear stress develops against the 'A' material and it, too, suffers from shortening. The net result of this mechanism is that, when the specimen has transformed, it is a little shorter and fatter than it should be.

This is a perfectly possible effect. It seems to me that, if there is any directionality in the segregation patterns, it is conceivable that this may cause distortion troubles during heat-treatment. This will, of course, apply principally to tool steels and carburizing steels, where one is interested in very close control of dimensions during heat-treatment involving transformations of this type.

AUTHORS' REPLY

Mr. **Thorneycroft**: Mr. Bygate's remarks reflect the considerable interest which he and his colleagues have always shown in the origin of low-hardenability zones in

steel ingots. In particular it is interesting to hear that in their further experiments on ingot material, they find no effect of some of the obvious variables, such as pouring speed, ingot size etc., on the occurrence of the tapered zone of low-hardenability.

The specific points raised by Mr. Bygate are replied to as follows. Segregation within the 9 s.w.g. wire was not thoroughly explored, and it is likely that larger differences in progress of transformation across the section would have been found in samples partially transformed at temperatures other than 550° C. That some segregation remained in the wires was shown more clearly by the progress of the martensite transformation in axial and near-surface zones. Nevertheless, by comparison with ingot material, the wires were relatively homogeneous.

As shown in Fig. 5, the pearlite reaction times for ingot material change rapidly over the temperature range 600–700° C, and this feature accounts for the differences in reaction times recorded in Tables IV and V.

We agree that neither ingot nor bar material was fully homogenized by the few days, treatment at 1200° C. The evidence of Figs. 17*b*, 18, and 19 betrays a residual heterogeneity, and it is probable that these last traces of segregation will be difficult to remove, since the concentration gradient, a driving force in diffusion, will be slight compared with that of the untreated sample.

The example of transverse cracking in an alloy steel billet parallels that encountered by the authors. The degree of microsegregation in the cracked, central zone of the billet was sufficient to produce a substantial difference in bainite reaction times between the centre and surface of the billet. It is probable that under conditions of continuous cooling, severe transformation stresses are engendered in the alloy-enriched interdendritic zones when these finally transform. As shown in the present paper, the difference in M_s temperatures, for example, between dendritic and interdendritic regions is 70° C. Such differential transformation in the martensite zone is likely to invoke the severest transformation stresses.

Mr. Gittus has described an interesting experiment in inoculation. The refinement of cast structure seems to be primarily in terms of reducing the length of the columnar crystal without change in width. Therefore, the distance for diffusion has not been radically altered. The reduction of coring was the salient feature, which warrants a more quantitative treatment.

We are delighted to hear from Mr. Kohn of the progress made in his studies of segregation. It would be useful to compare segregation intensity with the liquidus/solidus temperature intervals for the various binary and ternary alloys. This comparison may throw some light, for example, on the behaviour of the Fe–Co alloys. Incidentally, an autoradiographic technique is described in the April issue of the *Journal*, in which the radio-isotope ^{56}Mn was produced *in situ* by pile irradiation of ingot samples of 4%Ni–Cr steel. The segregation of manganese was confirmed.

It is possible that the pattern of microsegregation by Mr. Johnson's B.S. En 17 ingot was influenced by casting technique, or by a superimposed pattern of macrosegregation.

We agree with Mr. F. G. Haynes that banding in steel can be a source of distortion. Sheffield workers have shown examples of this effect, where, in the case of extreme strain between neighbouring bands, the stress level was high enough to produce localized transverse cracking.

CHAIRMAN'S CONCLUDING REMARKS

The **Chairman** (Dr. H. H. Burton): I should like to make two observations. First, there is evidently a very large series of problems still remaining for the Ingots Committee to take up. In the early days of the Ingot Committee, the causes of segregation were considered to be of paramount importance, but we have not solved them yet.

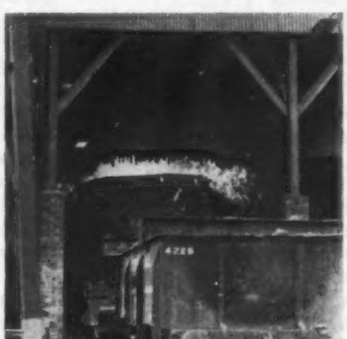
The second point is the frequent mention of banding. I suspect that some of those who have mentioned banding have not read very thoroughly the excellent paper that was given as the Hatfield Memorial Lecture in 1957. I suggest that they should read it, not once, but many times.

Tonnage Oxygen Plant at Scunthorpe

The Earl of Ancaster, T.D., officially opened the tonnage oxygen plant of British Oxygen Gases Ltd., at Scunthorpe on 7th October. The Company now has four tonnage plants; those at Margam and Scunthorpe are now in full production, and will shortly be augmented by further plants at Ebbw Vale and Consett.

The Scunthorpe plant has a daily capacity of 200 tons, which makes it the largest in commission in the U.K. It will serve the three iron and steelworks in the area—the Appleby-Frodingham Steel Company, John Lysaght's Scunthorpe Works Ltd., and the Redbourn Works of Richard Thomas and Baldwins Ltd.—to which it is linked by pipeline. It will have a considerable effect in increasing steel production in these works.

Oxygen now figures prominently in steelmaking practice in the Scunthorpe area. The "Ajax" furnace at Appleby-Frodingham (*J.I.S.I.*, 1958, Sept., vol. 190, pp. 1–29) represents an important innovation in steelmaking practice; this tilting open-hearth furnace, modified so as to permit a much heavier use of oxygen than ever before, has already produced over 100,000 tons of steel. In a recent four-week period, its output was 19,263 tons, which is believed to be the highest figure ever attained by an open-hearth furnace in the U.K.



Ladle desiliconization with oxygen at Richard Thomas and Baldwin Ltd., Redbourn Works

At Redbourn, oxygen has been used for steelmaking for many years. Early experiments with oxygen lancing in the open-hearth furnace were found to speed up refining times and increase outputs by 5–10%. Later, oxygen was used for desiliconizing molten iron in the ladle, which further reduced refining times and raised production; plans are now well advanced for the installation of the first plant of its kind in the U.K. using oxygen for the removal of phosphorus. Since the tonnage plant began production, all seven open-hearths at Redbourn have been modified to permit the use of oxygen.

The new plant is of the Rescol type, using regenerators for air purification and a single distillation column for air separation. Expansion turbines are also a feature of the plant, which is capable of producing both liquid and gaseous oxygen. The oxygen is supplied to the steelworks in gaseous form, but nearly 600 tons of liquid oxygen is stored at the plant, to maintain supplies during maintenance shutdowns and to supplement gas supplies during peak demand periods.

IRON AND STEELWORKS ENGINEERING

The Iron and Steel Engineers Group

The Vacuum-lifter Process of the Dortmund-Hörder Hüttenunion AG

By F. Harders

SYNOPSIS

About 40,000 tons, mainly of unalloyed open-hearth steel, have so far been treated with the vacuum-lifter process of the Dortmund-Hörder Hüttenunion. Defects due to hydrogen are not now observed, although the heat treatments formerly necessary to reduce hydrogen content are no longer carried out. Oxide inclusions have been reduced to less than a half of their former level. In the production of heavy forgings, the vacuum-lifter process saves five times as much as it costs. Contraction cavities in killed steels are welded up much more readily in rolling than in untreated steels. Hot-rolled strips from vacuum-treated killed steel are absolutely free of laminations. 1630

THE CHARACTER of all chemical reactions is essentially influenced by the concentration of the elements concerned, the pressure, and the temperature. By varying these three factors it is possible to find the most favourable conditions for any industrial process based on chemical reactions with respect to productivity, economy, and quality of the finished product. For instance, the chemical industry achieved great economies in the gas synthesis process by changing the pressure. The degassing of steel by vacuum treatment on a production scale was introduced much later into steelworks. Although the first aim was to eliminate the detrimental hydrogen content, interest soon centred upon the metallurgical reactions.

This fact is especially significant in relation to the reaction of carbon with oxygen to produce carbon monoxide. If the partial pressure of CO in a balanced steel melt is reduced from 760 to 7·6 mm Hg, i.e. to one-hundredth of the atmospheric pressure, the oxygen content of a steel melt containing 0·1%C is lowered from 0·025% to 0·0003%, resulting in a reduction of the carbon content to 0·08% (*see* Fig. 1). In a very short time 0·35 m³ (N.T.P.) of CO per ton of treated steel can be drawn off. In a steel melt containing 0·5%C and subjected to the same pressure reduction, the oxygen content is lowered from 0·005% to 0·00005%, whereas the change in carbon content is so small that it does not become apparent. In this case, 0·07 m³ (N.T.P.) of CO per ton of steel are drawn off. It is therefore evident that pressure reduction is a powerful method of lowering the oxygen content of a steel without rendering it impure by the addition of solid deoxidizing agents. The carbon content is also reduced, but the effect is greater with soft steels than with hard steels, as the above examples show.

It would naturally be very tempting to remove the entire oxygen content and its chemical compounds from the steel by vacuum treatment: it is in fact possible to attain very low oxygen contents in the laboratory. However, the cost of vacuum treatment of steel melts increases in proportion to the degree of purity attained. Vacuum melting is economical only with high-grade steel requiring expensive alloying additions. It offers the advantage that the melt is allowed to degass for long periods under a high vacuum. The application of vacuum melting to steel will be restricted until the vacuum equipment can be enlarged and produced at a lower cost. The detrimental effect of non-metallic inclusions depends on the use to which the steel is to be put, and so it is not always necessary to attain a high degree of purity. It would therefore serve a very useful purpose if an inexpensive method of varying the amount of degassing to achieve the required degree of purity could be developed. The development of production methods has been encouraged by the fact that with most mass-produced steels a simple vacuum treatment can achieve quality changes and economic advantages.

The economies of operating a large steelworks depend on the trouble-free throughput of large quantities of steel without interference with the normal flow of operations. Thus the less a steel degassing method interferes with the normal flow of operations in the melting shop, the casting bay, etc., the better it is. Shorter degassing times interfere less with the normal flow of operations. Reduction in the time of treatment is also an advantage in that the

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Dr. Harders is a Director of Dortmund-Hörder Hüttenunion AG.

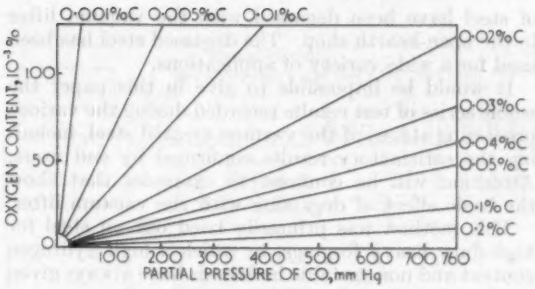


Fig. 1—Carbon-oxygen equilibrium as a function of partial pressure of CO

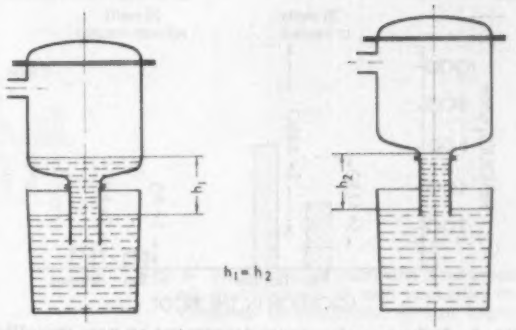


Fig. 3—Method of operation of vacuum lifter

detrimental effects of the degassed steel reacting with the refractory lining and the slag are reduced, since the gaseous CO can be more quickly removed from the melt than oxygen can be made available from the refractory lining or the slag, which only occurs in a liquid state.

Dortmund-Hörder Hüttenunion AG has developed a method for degassing mass-produced steels that takes all these factors into consideration. Figure 2 is a diagrammatic representation of the equipment used for degassing. A suction pipe dipped into the ladle draws steel off in small quantities and forces it into an evacuated vessel—the vacuum lifter. The degassed steel then flows back into the ladle. The vacuum lifter is continuously lowered and lifted, filling with steel and discharging it in a degassed condition into the ladle. Figure 3 shows the cycle of operations; the lower position is on the left and the upper position on the right. The difference in level of molten steel in the ladle and in the vacuum lifter is always the same, and is determined by barometric pressure. In the upper position the whole barometric steel column is in the suction pipe and no steel flows into the vacuum vessel itself. If the lifter is lowered without changing the pressure, the steel will flow into the vessel so long as the same barometric pressure is maintained. Lifting and lowering is repeated continuously until the desired degree of purity is achieved for the total content of the ladle. The vacuum lifter is electrically heated to counteract the cooling of the steel during vacuum treatment. This also keeps the temperature in the vacuum lifter, even during the intervals between the treatment cycle, so high that the efficiency of the refractory lining is not impaired by a change of temperature. Vigorous gas emission in the vacuum lifter allows the oxygen and hydrogen contents of the steel to be reduced to values that cannot be achieved under atmospheric pressure. The steel is then teemed in the normal way or under a screen of protective gas. By introducing the degassing

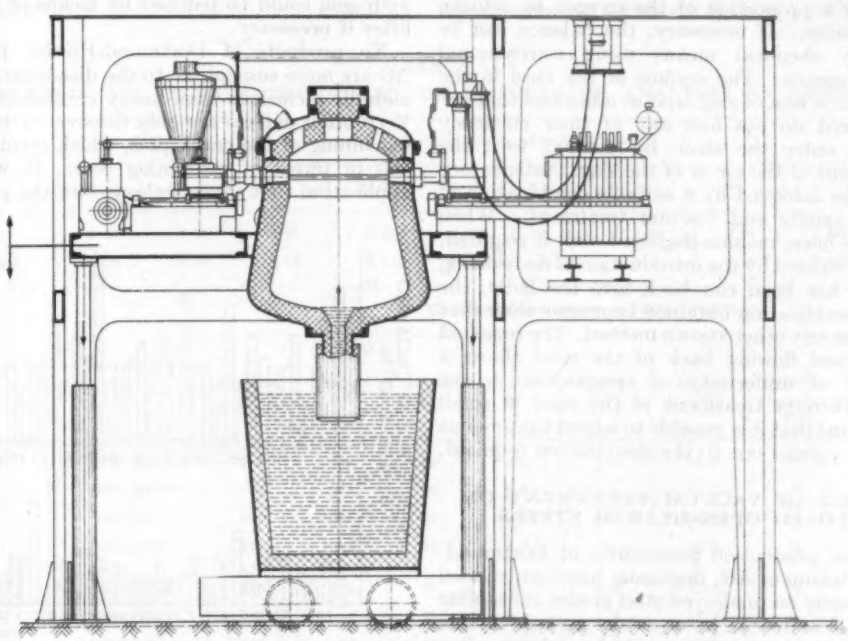


Fig. 2—Vacuum-lifter plant

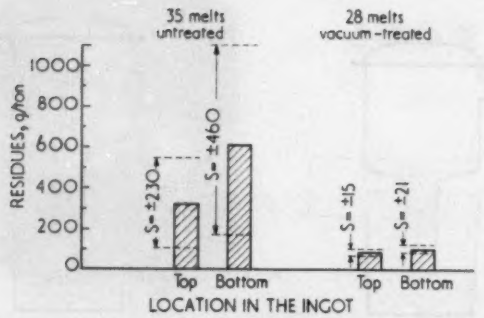


Fig. 4—Influence of vacuum treatment on non-metallic inclusions. Unalloyed 45-ton ingot containing 0·22–0·45%C. *S* = standard deviation

process between tapping and pouring, this method complies with the requirement mentioned earlier that a production-scale degassing process must not interfere with the normal flow of operations in the steelworks and must also be independent of the ingot size.

It is possible to vary the concentration of materials, the pressure, and the temperature in the vacuum lifter within certain limits. Steel quality can be influenced in many ways and with graded effects. The effect of such variations is so pronounced that it is impossible at present to predict the range of application of the degassing method. However, the operational results achieved so far with the vacuum lifter prove that this method can profitably be used for numerous purposes. Although it is not yet possible to supply a vacuum lifter for each steel quality, complete with operating instructions, the results so far achieved show that it is worthwhile investigating the possibility. In many cases it will be sufficient to remove only a proportion of the oxygen to achieve excellent results. If necessary, the balance can be removed by chemical means using conventional deoxidizing agents. The surface of the steel in the vacuum lifter is free of slag, and so additions that are easily oxidized do not lose any of their efficiency before they enter the steel, by contact with the oxygen content of the air or of the slag. Satisfactory results can be achieved by a suitable combination of deoxidizing agents and vacuum treatment. When the steel has been vacuum-degassed and, if required, additionally treated by the introduction of deoxidizing agents, and has been run back into the ladle, the degree of deoxidization obtained is greater than that resulting from any other known method. The repeated drawing-off and flowing back of the steel allows a high degree of uniformity of composition to be attained. Vacuum treatment of the steel in small amounts means that it is possible to adjust the amount of degassing carried out to the deoxidation required.

INFLUENCE OF VACUUM TREATMENT ON UNALLOYED OPEN-HEARTH STEELS

To suit the production programme of Dortmund-Hörder Hüttenunion AG, degassing has been carried out so far chiefly on unalloyed steel grades containing 0·05–0·65%C and to a lesser extent on free-cutting steel, spring steel, manganese steel, and low-alloy chromium steel. Up to the present about 40,000 tons of steel have been degassed with the vacuum lifter in the open-hearth shop. The degassed steel has been used for a wide variety of applications.

It would be impossible to give in this paper the whole series of test results recorded during the various processing stages of the vacuum-treated steel, including the satisfactory results confirmed by end users. Attention will be confined to examples that show the basic effect of degassing with the vacuum lifter.

This method was primarily tried out on steel for high-duty heavy forgings, in which a high hydrogen content and non-metallic inclusions have always given rise to a large number of rejects.

Hydrogen is particularly detrimental in alloy steels, but it is of minor importance in forgings of unalloyed steel produced by Dortmund Hörder-Hüttenunion AG. Nevertheless, before the installation of the vacuum lifter the works occasionally discovered defects due to the hydrogen content. Since the introduction of degassing there have been no indications of defects from this cause. Moreover, for more than a year the works have been able to dispense with the long-time process annealing previously needed to reduce the hydrogen content in untreated steel. The slow cooling rate is no longer necessary. Heavy forgings produced from unalloyed vacuum-treated steel can be left to air-cool without danger. These facts are better proof of the influence of the vacuum lifter on the hydrogen content of unalloyed steels than figures that depend on the uncertain factor of the method of sampling. Experience so far has shown that it will be sufficient so far as the production programme of Dortmund-Hörder Hüttenunion AG is concerned if the hydrogen content is reduced to half the original value. For this reason, there have been no scheduled tests to find out the extent to which hydrogen could be reduced by means of the vacuum lifter if necessary.

No products of Dortmund-Hörder Hüttenunion AG are more susceptible to the disadvantages of non-metallic inclusions than heavy crankshafts for ships. Very often defects are only discovered during finish-machining of the crank pins, which involves a heavy loss in terms of machining cost. It was for this application that investigations into the possibility of

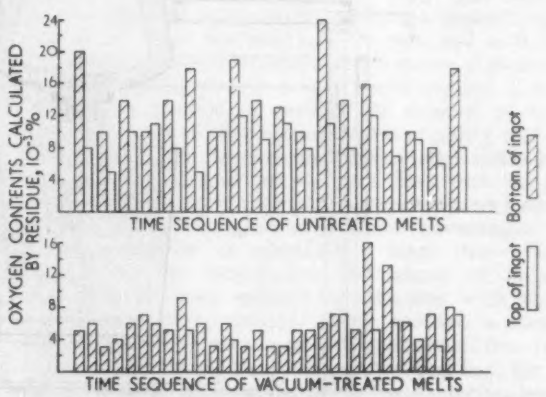


Fig. 5—Influence of vacuum treatment on oxygen content. Unalloyed 45-ton ingots containing 0·35%C

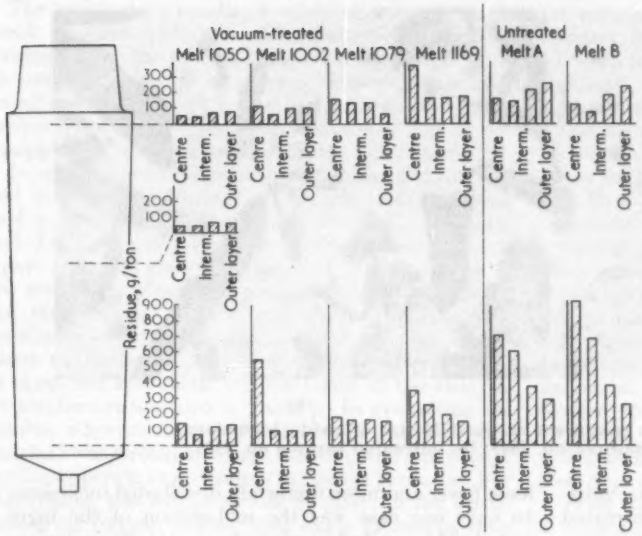


Fig. 6—Influence of vacuum treatment on distribution of non-metallic inclusions in unalloyed 45-ton ingots

degassing steel melts began, to find out how much the purity of the steel could be increased.

Operating conditions at the plants of Dortmund-Hörder Hüttenunion AG for producing sound crankshafts were particularly unfavourable before the installation of the vacuum lifter. The steel had to be refined in 150-ton open-hearth furnaces equipped for mass production; this meant that production of crankshafts was unprofitable. Long before the vacuum lifter was installed, the high percentage of rejects had made it necessary to buy forging-grade ingots for crankshafts from other steelworks producing high-grade steel.

Only in this way could the rejects be sufficiently reduced to permit the production of crankshafts. But the economics of operation were still not satisfactory. Only when the vacuum lifter was installed was improvement in quality of forging-grade ingots possible to the required degree and steel made in the normal melting shop could be used economically.

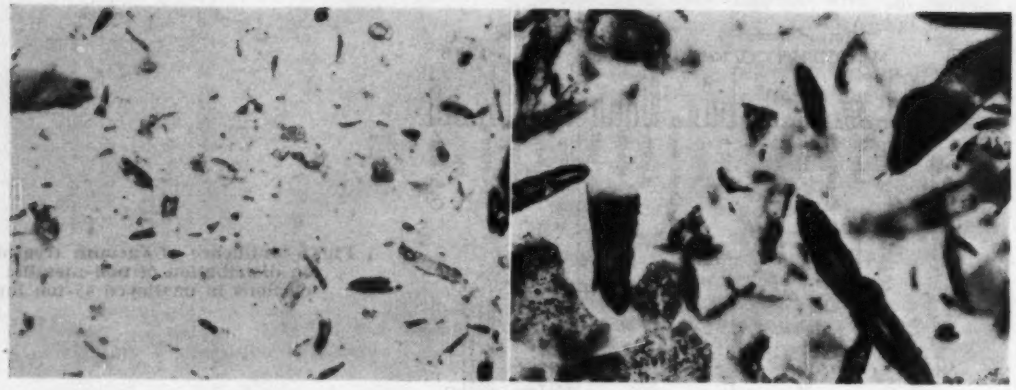
Crankshafts produced from untreated steel from the melting shop have not been made for so long that the crankshafts now produced from vacuum-treated steel can only be compared with those made by other companies. The untreated forging-grade steels described below are therefore those produced by other

Table I

INFLUENCE OF VACUUM TREATMENT ON RESULTS OF ANALYSIS BY RESIDUE

	Chemical Composition of Steel									
	C, %	Si, %	Mn, %	P, %	S, %	Cu, %	Cr, %	Mo, %	Al, %	N_2, %
Untreated	0·15	0·31	1·07	0·03	0·016	0·13	1·01	0·09	0·005	0·003
Vacuum-treated	0·14	0·32	1·08	0·029	0·017	0·13	1·00	0·09	0·005	0·003

	Sample No.	Chemical Composition of Oxide Residue							
		Residue, g/l.	SiO_2, g/l.	Al_2O_3, g/l.	FeO, g/l.	MnO, g/l.	Cr_2O_3, g/l.	O_2	
								Determined, 10^{-3} %	Calculated, 10^{-4} %
Untreated	2350 BA/R	212	148	39	tr.	0·4	18·7	12 11	10
	2351 BA/M	194	119	28	5·0	2·1	19·0	9 10	9
	2352 B/R	220	127	33	6·6	tr.	30·6	11 11	10
	2353 B/M	343	208	61	4·8	tr.	29·4	15 23	16
Vacuum-treated	2346 BA/R	148	101	20	tr.	3·3	12·1	6	7
	2347 BA/M	145	98	25	0·9	1·3	11·7	10	7
	2348 B/R	109	73	17	0·3	0·2	10·0	9 8	5
	2349 B/M	105	68	18	0·7	1·7	8·8	7	5



(a) (b)

Fig. 7—Influence of vacuum treatment on particle size of oxide inclusions. (*a*) Vacuum-treated sample no. 2347; (*b*) untreated sample no. 2351

companies. This also explains why different deoxidizing agents have been used with treated and untreated steel.

Figure 4 shows the effect of vacuum treatment on the amount and distribution of non-metallic inclusions in crankshafts. The top and bottom parts of the ingots were the crankshaft ends from which the test specimens were taken. The tests covered 35 ingots produced from untreated steel and 28 ingots produced from vacuum-treated melts containing 0·22–0·45%C; ingot weight was in all cases 45 tons. The untreated melts were deoxidized in the ladle by the addition of calcium–silicon (CaSi). Ferro-silicon (FeSi) was added under vacuum to the vacuum-treated melts after extensive degassing. Test specimens taken from the steel were electrolytically treated to determine the degree of purity. Figure 4 shows the residues of the electrolytic treatment in grammes per ton of steel. Treatment with the vacuum lifter allowed a large reduction in non-metallic inclusions, and furthermore the differences between top and bottom became of minor importance.

Figure 5 shows the results of a further series of tests on heavy crankshafts. The non-metallic inclusions differ from ingot to ingot. These tests only concerned crankshafts produced from unalloyed 45-ton forging-grade ingots containing 0·35%C, of which 18 were produced from untreated and 27 from vacuum-treated melts. Test specimens were taken in the same way as those described in Fig. 4. It was not thought necessary to test all the top ends of the vacuum-treated ingots, because most of the non-metallic inclusions are generally contained in the bottom ends. In spite of some variations, which are so insignificant that they may be attributed to differences in teeming conditions, oxide inclusions in crankshafts produced from vacuum-treated steel are only about 40% of those in crankshafts from untreated steel melts.

Figure 6 shows the distribution of non-metallic inclusions over the cross-section of the ingots. The amount of inclusions is expressed in grammes per ton of steel resulting from electrolytic treatment of test specimens taken from the core, intermediate layer, and outer layer of some 45-ton ingots. The bottom part of ingots of vacuum-treated steel in particular have fewer and more uniformly distributed inclusions. In only one case was the mid-section of the ingot tested. The smallest number of non-metallic inclusions were discovered in this part of the ingot.

About 5500 tons of vacuum-treated steel have to date been processed into 121 crankshafts, of which 80 have already been finish-machined and dispatched; 15% were rejected when finished. This is a considerable improvement in that the Company's earlier net output was so bad that the figures were never published. If anyone intends to make a comparison, he should find out for himself the situation obtaining in his own plant. It should be borne in mind that Dortmund-Hörder Hüttenunion AG is primarily a mass-production steel company.

Another advantage which the vacuum lifter offers from the forging standpoint is the better closing of secondary pipe. This phenomenon can better be followed up during rolling, and so a detailed description is given later in the section dealing with rolled products.

Many and extensive tests will be necessary before quantitative data can be given about changes caused by the vacuum lifter in high-duty heavy forgings, but there is already clear evidence of the advantages it can offer. For example, it is economical for the forging shop to pay for each ton of vacuum-treated steel an additional amount equivalent to five times the maximum cost of the vacuum treatment. It must be taken into consideration, moreover, that the vacuum lifter referred to in this paper is the first installation of its kind, and has not yet achieved its optimum efficiency.

Table II

SECTIONS WITH PIPE OBSERVED DURING ULTRASONIC TESTING

	Slab Ingot I	Slab II	Plate III	Hot-rolled Strip IV
Untreated	100%	35%	15%	1·5%
Vacuum-treated	100%	20%	<5%	0

The satisfactory results of applying vacuum treatment to steel melts for the production of heavy forgings led to the process being tried on a larger scale on steels intended for other uses. However, the tests on alloy steels had to be limited to low-alloy steels because Dortmund Hörder Hüttenunion AG is not equipped for the production of high-alloy steels.

In order to get an idea of the influence of chromium and manganese additions on the degassing process, half a 150-ton melt containing 0·15%C, 1·0%Mn, and 1·0%Cr was subjected to vacuum treatment. Apart from the vacuum treatment, both portions of the melt were given the same treatment as regards the metallurgical process and shaping. Four test specimens were taken from identical locations in the pieces at the same stage of hot-working. The analysis of these test specimens and the result of the analysis by residue are included in Table I. In evaluating the results, allowances must be made for the fact that the steel was melted in a large furnace, which naturally does not operate so accurately as smaller furnaces. For this reason, less weight should be given to the actual values for oxygen content and oxide inclusions than the fact that treatment with the vacuum lifter allowed a reduction of about 50% of the total oxide residues. The composition of the residues and the steel was not appreciably changed. Only the FeO and MnO contents deviate from the average. But the percentage of these components in the residue and the number of test values are too small to justify any inference from these observations. The large reduction of the oxide residue is no less impressive than the change of average size of the single particles shown in Fig. 7.

It is remarkable that no aluminium was added in the ladle to the vacuum-treated steel or to the untreated melt.

It has already been mentioned that the closing of secondary pipes is easier in vacuum-treated than in untreated steel. This phenomenon, which is of utmost importance for the net output, has been investigated on samples of 40 slab ingots rolled into plates and sheets. The weight of the slab ingots ranged from 14 to 21 tons, and they had been top-poured without sinkhead. The steels had been deoxidized by adding aluminium. The slab ingots, slabs, plates, and hot-rolled strip were subjected to ultrasonic testing or inspection of the sectional areas. The test results are given in Table II. The tops of all the slab ingots had pipe formed during teeming; the size was not influenced by vacuum treatment. The slab ingots of both melts were rolled into similar-sized slabs. With the untreated steel, 35% of the slabs still contained shrinkage pipe, whereas with the vacuum-treated steel this proportion was only 20%. Rolling the slabs into heavy plates led to a further closing of the remaining shrinkage pipes, which decreased to 15% with the untreated and to less than 5% with the vacuum-treated steel. After rolling the slabs into hot-rolled strips the untreated steel generally still shows about 1·5% shrinkage pipe, whereas in the vacuum-treated steel this was in all cases welded together.

After the steel has been rolled into small sections, a large number of specimens can be taken without difficulty, which explains why the investigation has covered such a very wide range, since, in addition to forgings, rolled products have been made from vacuum-treated steel to an increasing extent. These comparative investigations of vacuum-treated and untreated steels, having been designed on a large scale and supported by fundamental research and careful determination of the behaviour in service, have led to valuable results. However, it is not proposed to make a detailed statement on this, since the interpretation of all the new data is not complete. The small number of results reported, however, should be sufficient to prove that with the vacuum-lifter process the degree of purity and the gas content of plain carbon and low-alloy steels can be controlled. Furthermore, the example of the heavy crankshafts shows that the use of the vacuum lifter can prove to be of economic benefit.

Sea-Water Magnesia Plant

The 1958 Meeting of the Refractories Association of Great Britain was held at Scarborough in September. During the course of the Meeting, members were invited by the President, Mr. Neil M. Peech, to visit the sea-water magnesia works of the Steetley Company Ltd., of which he is Chairman.

The present 24-acre coastal site was acquired in 1937, and a pilot-plant was installed; Mr. Peech had recognized the potentialities of the process, and it was on his initiative that this plant was built, despite the scepticism of many people. He was encouraged in the venture by Mr. R. F. Summers and by Sir Andrew Duncan. In 1938 a full-scale plant was put down, the first of its kind in the world, with a scheduled annual capacity of 10,000 tons. There were many teething troubles to be overcome, but by 1939 production had reached 4,000 tons.

During the war the demand for magnesia increased tremendously, since it was a major source of magnesium metal, and the capacity of the plant was repeatedly increased. By 1952 output had reached 54,000 tons annually, and this has now been raised to 150,000 tons in 1958. The bulk of the production is now supplied to the refractories industry. The British industry is now no longer dependent on Austrian magnesite, and the cost of *Britmag*, the trade name for the sea-water magnesia, is low enough for it to compete with natural magnesite in overseas markets.

One of the three settling tanks at the Hartlepools sea-water magnesia plant

N E W S

Announcements and News of Science and Industry

THE IRON AND STEEL INSTITUTE

Autumn General Meeting

The Autumn General Meeting is being held in London on **10th December;** the 11th Hatfield Memorial Lecture and the 38th Meeting of the Iron and Steel Engineers Group will also take place in conjunction with the Meeting. The full programme is as follows:

Tuesday, 9th December

6.00 p.m. Weir Lecture Hall, Institution of Naval Architects, 10 Upper Belgrave Street, London S.W.1 (tickets NOT required).
11th Hatfield Memorial Lecture: "The Mechanical Properties of the Ferrite Crystal," by Dr. N. P. Allen, F.R.S.

Wednesday, 10th December

Autumn General Meeting

Morning Session

(at 4 Grosvenor Gardens, London S.W.1)

10.0–10.15 a.m. Formal business and election of Members.

Session A—Continuous Casting of Steel
(at 4 Grosvenor Gardens, London, S.W.1)

10.30 a.m. Introductory talk and presentation of the following papers (discussed at Charleroi during the Journées Internationales de Sidérurgie) by Mr. G. N. F. Wingate (Distington Engineering Co., Ltd.):
- "Continuous Casting on the B.I.S.R.A. Experimental Plant," *by G. Fenton and J. Pearson* (*Journal*, June, 1958)
- "Continuous Casting at the Centre National de Recherches Métallurgiques," *by J. Zaeytydt* (October, 1958)
- "Control of Flow of Liquid Metal during Continuous Casting," *by W. Siegfried and B. Broniewski* (October, 1958)
- "Continuous Casting of Steel in France," *by P. Thomas* (October, 1958)
- "Continuous Casting of Steel in Western Germany," *by K. G. Speith and A. Bungeroth* (October, 1958)
- "New Experience in the Continuous Casting of Steel," *by H. Krainer and B. Tarmann* (October, 1958)
- "Continuous Casting at Atlas Steels Ltd., Welland, Ontario, Canada," *by G. C. Olson* (September, 1958)

10.45 a.m. Showing of films on continuous casting at Cail, Atlas, Jessops, and Barrow.

12 noon–1.15 p.m. Presentation and discussion of the following papers (both to be preprinted):
- "Continuous Casting at Barrow," *by I. M. D. Halliday.*
- "Continuous Casting of Steel in the U.S.S.R.," *by M. S. Boichenko, V. S. Rutes, D. P. Evteev, and B. N. Katomin.*

Session B—Creep in Steels
(at the Institution of Naval Architects, 10 Upper Belgrave Street, London, S.W.1)

10.30 a.m.–1.15 p.m. Presentation and discussion of the following papers:
- "A New Approach to the Problem of Creep," *by J. Glen* (August, 1958)
- "The Effect of Alloying Elements on Creep Behaviour," *by J. Glen* (October, 1958)
- "Ductility in High-temperature Rupture Tests," *by J. Glen* (September, 1958)
- "Recent Developments in Creep Testing by the Cantilever Bending Method," *by G. T. Harris, H. C. Child, A. B. Collier, and C. F. West* (October, 1958)

Afternoon Session

2.30–5.30 p.m. Two simultaneous technical sessions:

Session A (contd.)—Continuous Casting of Steel (at 4 Grosvenor Gardens, London, S.W.1)

2.30 p.m. Showing of film from B.I.S.R.A., introduced by Dr. J. Pearson.

3 p.m. Presentation of the following papers:
- "Solidification Characteristics of Continuously Cast Low-carbon Steel," *by C. F. Schrader, M. Tenenbaum, and L. Mair* (to be preprinted)
- "Continuous Casting of High-speed Steel," *by B. H. C. Waters, W. H. Pritchard, A. Braybrook, and G. T. Harris* (November, 1958)

4.0–5.30 p.m. General discussion on continuous casting

Session C—Hydrogen in Weld Metal (Joint Meeting with The Institute of Welding) (at the Institution of Naval Architects, 10 Upper Belgrave Street, London, S.W.1.)

2.30–5.30 p.m. Presentation and discussion of the following papers (all published in the *British Welding Journal*):
- "Hydrogen Content of Mild and Alloy Steel Weld Deposits," *by P. D. Blake* (March, 1957)
- "Measurements on the Diffusible, Residual, and Total Hydrogen Contents of Weld Metal," *by P. D. Blake* (March, 1958)
- "Effect of Diffusible Hydrogen on the Mechanical Properties of Weld Metal," *by P. D. Blake* (December, 1958)
- "Hydrogen in Mild-steel Weld Deposits," *by N. Christensen and R. Rose* (December, 1955)
- "Hydrogen in Mild-steel Weld Deposits," *by N. Christensen, K. Gjermundsen, and R. Rose* (June, 1958)
- "Effects of Hydrogen in Wrought Steel and Ferrous Weld Metal," *by P. D. Blake and W. I. Pumphrey* (to be preprinted)

Thursday, 11th December

38th Meeting of the Iron and Steel Engineers Group on "Vacuum Processes in the Iron and Steel Industry" (at 4 Grosvenor Gardens, London, S.W.1).

Morning Session

10 a.m.–1 p.m. Presentation and discussion of the following papers:
- "Vacuum Degassing of Steel with the Vacuum Casting Method of Bochumer Verein," *by G. Bandel*
- "Vacuum Pouring of Ingots for Heavy Forgings," *by J. H. Stoll*
- "The Vacuum-lifter Process of Dortmund-Hörder Hüttenunion AG," *by F. Harders*
- "Vacuum Casting of Heavy Forging Ingots," *by J. Levaux and L. Hanon*

"Equipment used for Vacuum Melting and Casting in the Iron and Steel Industry," *by A. S. D. Barrett and M. Harper*

Afternoon Session

2.30–4.30 p.m. Presentation and discussion of the following papers:

"Vacuum Induction Melting at Utica," *by F. N. Darmara*

"Vacuum Melting of Steels," *by H. C. Child and G. T. Harris*

"The Behaviour of Various Elements in Vacuum Steel Melting," *by G. M. Gill et al.*

Preprints of these papers will be sent to all Members attending the Engineers' Group meeting.

NEWS OF MEMBERS

Dr. **N. P. Allen,** F.R.S., has been elected a Vice-President of The Institution of Metallurgists.

Dr. **W. A. Baker** has left the British Non-Ferrous Metals Research Association to join the Northern Aluminium Co., Ltd.

Mr. **W. E. Ballard** has been elected President of The Institution of Metallurgists for 1958–1959.

Mr. **N. R. R. Brooke** is a member of the Board of the Development Corporation for Wales.

Mr. **A. Cibula** has been appointed Joint Research Superintendent of the British Non-Ferrous Metals Research Association.

Mr. **W. J. G. Cosgrave** is now Metallurgist to the Midland Sales Area for the Electric Furnace Co., Ltd., and the Electric Resistance Furnace Co., Ltd.

Mr. **J. T. W. Dewar** is now Managing Director of Firth-Vickers Stainless Steels Ltd., following the retirement of Mr. C. E. Holmstrom.

Mr. **R. Eborall** has been appointed Joint Research Superintendent of the British Non-Ferrous Metals Research Association.

Dr. **T. P. Hoar** is to receive the Whitney Award of the National Association of Corrosion Engineers for 1957.

Mr. **C. E. Holmstrom** retired on 28th July from the post of Joint Managing Director of Firth-Vickers Stainless Steels Ltd., after 46 years with the Company and its predecessors. He has been appointed a non-executive Director.

Dr. **D. H. Houseman** has been appointed Head of the Metallurgy Department of Sir John Cass College.

Sir **Walter Benton Jones,** BT., has been appointed a member of the new governing body of Ashridge College, Berkhamsted.

Dr. Ing. **J. Koritta** is now Professor at Prague University.

Mr. **G. Meikle** has been elected a Vice-President of The Institution of Metallurgists.

Mr. **U. le Paige** is now Directeur Technique of the S.A. Métallurgique d'Espérance-Longdoz.

Mr. **E. J. Pode** is Deputy Chairman of the Development Corporation for Wales.

Mr. **E. Ross** has joined the Research and Development Branch of the U.K. Atomic Energy Authority as a Senior Scientific Officer.

Mr. **M. N. Saxena** is now Reader in Mechanical Engineering (Metallurgy) at the University of Rourkee, India; he recently left for the U.S.A. to work for his research degree in the Department of Mining and Metallurgy at the University of Wisconsin.

Mr. **D. R. Sen** is now Assistant to the Technical Industrial Management Department and Production Control at Klöcknerhütte Bremen AG.

Mr. **H. J. Sharp** has been awarded the degree of Ph.D. by London University.

Mr. **A. Smithson** has been appointed General Sales Manager of Hartons Installations Ltd.

Mr. **H. F. Spencer** is a member of the Board of the Development Corporation for Wales.

Mr. **F. W. E. Spies** has been promoted to Assistant Director of the Royal Dutch Blast Furnaces and Steelworks Ltd., IJmuiden.

Sir **Geoffrey Summers,** BT., C.B.E., is a member of the Board of the Development Corporation for Wales.

Mr. **C. R. Wheeler,** C.B.E. (*President*), is a member of the Board of the Development Corporation for Wales.

Mr. **F. Wild** has been appointed Section Leader of the Mechanical Test Section in the Metallurgy Department, Research Division, of A. V. Roe and Co., Ltd.

Mr. **F. S. Wilson** has joined the English Steel Corporation Ltd., as a metallurgist in the foundry.

Obituary

Mr. **John Gray Buchanan,** J.P. (*elected* 1937), of London, on 22nd April, 1958.

Mr. **Thomas Wilton Hand,** M.I.MECH.E. (*elected* 1917), of Hathersage, Derbyshire, on 29th August, 1958.

Mr. **George Husband,** A.M.I.MECH.E. (*elected* 1953), of Stockton-on-Tees, on 24th September, 1958.

Mr. **Frederick Hamilton Keating** (*elected* 1935), of Eaglescliffe, Co. Durham, on 16th August, 1958.

Mr. **Charles Benjamin Lang** (*elected* 1938), of Montreal, Canada, in February, 1958.

Mr. **Frederick A. McCoy** (*elected* 1951), of Kansas City, Mo., U.S.A., on 25th August, 1958.

Mr. **Arthur Poole,** M.I.MECH.E. (*elected* 1933), of Sheffield, on 16th July, 1958.

Mr. **Francis Bowyer Rice-Oxley** (*elected* 1936), of London, on 18th August, 1958.

AFFILIATED LOCAL SOCIETIES

East Midlands Metallurgical Society

The Officers and Committee of the Society for the 1958–1959 session are:

President

N. S. Angus, A.MET., A.I.M.

Vice-President

H. E. Gresham, B.SC., A.R.I.C.

Treasurer

E. D. Knights, B.SC., F.I.M.

Secretary

P. A. Blackwell, B.SC., A.I.M.

Committee

A. R. Bingham, A.I.M.
E. Darley, A.MET.
H. K. Lloyd, PH.D., A.I.M.
M. McIntyre, B.SC., A.I.M.
H. L. Parker, A.R.I.C.
T. Parry, F.I.M.

Sheffield Society of Engineers and Metallurgists

The Honorary Secretary of the Society is now Dr. J. F. Wallace, Department of Mechanical Engineering, The University, St. George's Square, Sheffield 1.

Swansea and District Metallurgical Society

The Officers and Council of the Society for the 1958–1959 session are:

President

R. Higgins, O.B.E., PH.D., B.SC., F.I.M.

Vice-Presidents

S. S. Carlisle, M.SC.
D. W. Hopkins, M.SC., F.I.M.

Honorary Treasurer

L. A. S. Perrett, F.I.M.

Joint Honorary Secretaries

M. J. A. Thomas, B.SC.
P. E. Morgan, L.I.M.

Council

H. A. Cooke (Chairman)
C. H. Cunniffe, C.B.E.
E. McVicar
P. Mackay
D. J. D. Unwin, B.SC., F.R.I.C., A.I.M.
R. L. Williams
D. A. Bishop, A.I.M.
R. H. Pullen

J. R. Brown

R. F. Hellon

CONTRIBUTORS TO THE JOURNAL

J. R. Brown, B.A. (HONS.)—Head of Metal Physics Section, G.K.N. Group Research Laboratory (*author of paper in October issue*).

Mr. Brown was educated at Wellington Grammar School and St. John's College, Oxford, where he read physics, obtaining the degree of B.A.(Hons.) in 1948. He then entered the Research Department of the Imperial Smelting Corporation, and worked there until 1951, when he joined his present organization. Mr. Brown has been working chiefly on X-ray diffraction, preferred orientation in silicon–iron, and thermodynamics of metals.

R. F. Hellon, D.PHIL., B.SC. (HONS.)—Climate and Working Efficiency Unit, Medical Research Council, Oxford (*author of paper in October issue*).

Dr. Hellon was educated at Merchant Taylors' School, Crosby, and Liverpool University, where he obtained a B.Sc. (Hons.) degree in physiology in 1949. After two years at the Clothing and Equipment Physiological Research Establishment (Ministry of Supply), he joined the scientific staff of the Medical Research Council, to work with the Climate and Working Efficiency Unit in the Department of Human Anatomy, Oxford. He obtained the degree of D.Phil. in 1957. Dr. Hellon's work has been concerned mainly with the effects of high temperature on man, and in particular with acclimatization and the effects of age on reactions to heat.

THE INSTITUTE OF METALS

Spring Meeting, 1959

The Institute's Spring Meeting will be held in London on Tuesday, Wednesday, and Thursday, 14th–16th April, 1959. The Annual Dinner–Dance will take place on the evening of Wednesday, 15th April.

Election of Honorary Members

The following elections to Honorary Membership of the Institute were announced recently:

Sir John Cockcroft, O.M., K.C.B., F.R.S.
Sir Christopher Hinton, K.B.E., F.R.S.
Professor Dr. phil. Dr.-Ing. E.h. Werner Köster.

In addition, Professor Dr. Ing. Alfred von Zeerleder has been elected a Fellow of the Institute.

Beilby Memorial Awards

Two awards, each of one hundred and fifty guineas, have been made by the administrators of the Sir George Beilby Memorial Fund, to:

Mr. B. E. Hopkins, M.SC., in recognition of his work in physical metallurgy.

Dr. E. C. Potter, in recognition of his work on electrochemistry and the corrosion of metals.

BRITISH NUCLEAR ENERGY CONFERENCE

Forthcoming Meetings

The following meetings have been arranged by constituent societies for the 1958–1959 session:

13th Nov.—"Electrification of U.K.A.E.A. Industrial Group Factories," by J. W. Binns and W. G. Outram—*Institution of Electrical Engineers*, Savoy Place, London, W.C.2, 5.30 p.m.

26th Nov.—"Recent Advances in Nuclear Engineering" (*Sir Charles Parsons Memorial Lecture*), by Sir John Cockcroft—*Institute of Marine Engineers*, 76 Mark Lane, London, E.C.3, 5.30 p.m.

4th Dec.—"Thermonuclear Processes," by D. W. Fry—*Institution of Electrical Engineers*, 5.30 p.m.

5th Dec.—"An Experimental Investigation into the Stress Distribution in a Band-reinforced Pressure Vessel," by K. G. Mantle, N. Marshall, and P. J. Palmer—*Institution of Mechanical Engineers*, 1 Birdcage Walk, London, S.W.1, 6 p.m.

15th Jan.—"Electrical Installation at Calder Hall Power Station," by N. J. Mackay and E. Hardwick—*Institution of Electrical Engineers*, 5.30 p.m.

22nd–23rd Jan.—Two-day Symposium on Nuclear Fuel Cycles, sponsored by *The Institute of Physics*.

28th Jan.—"The Design of Electro-mechanical Auxiliaries directly associated with Power-producing Reactors," by A. E. Harwood *et al.*—*Institution of Electrical Engineers*, 5.30 p.m.

11th Mar.—All-day Lecture Symposium on the 2nd International Conference on the Peaceful Uses of Atomic Energy (details to be announced later).

EDUCATION

School of Welding Technology

Following the success of the first year of its School of Welding Technology, when nearly 400 engineers attended courses, the Institute of Welding is offering a wider range of courses for the 1958–1959 Session. Courses on welding in shipbuilding, inspection and testing, training of welders, brazing technology and design, welded storage tanks, metal spraying, and welded design and construction in corrosion- and heat-resisting materials will be held in London, and a course on the detail design of welded structural steelwork is being run in Glasgow.

Prospectuses and enrolment forms may be obtained from the Secretary, The Institute of Welding, 54 Princes Gate, London, S.W.7.

"The Cast Structure in Metal Founding"

A course under the above title dealing with theoretical and practical aspects of structure in castings is being held from 15th to 17th December at the Department of Industrial Metallurgy, the University, Edgbaston, Birmingham 15 (fee two guineas).

Expansion of Technical Education

The Government's plans for the expansion of technical education are outlined in a pamphlet, "Britain's Future and Technical Education," published recently by the Ministry of Education, with the full support of the Federation of British Industries, the Trades Union Congress, the British Employers' Federation, and the Association of British Chambers of Commerce. The pamphlet describes the pattern being followed in the development of technical colleges in England and Wales, and shows how industry and commerce can play their part in the nation-wide scheme.

In a foreword, Mr. Geoffrey Lloyd, M.P., Minister of

Education, stresses the fact that the scheme can only succeed if it has the full backing of industry; without this, "we cannot be sure of having courses of the right type, or the students to put in them, or teachers to man them." The immediate future offers a great challenge, since the "bulge" will shortly be leaving school to enter the labour market; if steps are taken to deal with this group in good time, it can be of immense value in remedying the present shortages of skilled manpower.

NEWS OF SCIENCE AND INDUSTRY

Film on Gear Manufacture

Shell-Mex and B.P. Ltd. have produced a new film entitled "An Introduction to Modern Gears," continuing its policy of making instructional films for use in technological training. The present film is especially welcome, since there is no other general film on gear manufacture available. The technical advice of Dr. H. E. Merritt ensured that it would be authoritative in its field, and the co-operation of many leading manufacturing companies helped to make it as comprehensive as possible.

The film takes the form of a lecture, and is intended for use in schools, technical institutes, and industrial training centres. It begins with a recapitulation of the main functions of gears, and then classifies them into two major groups. The principles of gear design are discussed, and techniques for cutting and generating gear teeth are detailed. Methods used in research and testing are described, and the film ends with some contrasting examples of gear trains in present-day use.

The film is in black-and-white, and runs for 31 minutes; copies are available in both 35- and 16-mm gauge.

Electric Steel Makers' Guild

The Electric Steel Makers' Guild was formed recently to promote progress in electric furnace steel manufacture. Membership is limited to executives who are regularly employed in the manufacture of steel in electric furnaces for commercial use. Informal meetings are held three times a year. The President is Mr. F. T. Bagnall (Samuel Fox and Co., Ltd.) and the Vice-President Mr. A. Mettam (English Steel Corporation). The Secretary-Treasurer, from whom more details can be obtained, is Mr. W. H. Everard, Edgar Allen and Co., Ltd., Imperial Steel Works, Sheffield 9.

"Steel—Key to Britain's Future"

This is the title of a series of articles published recently by the *News Chronicle* and edited by Sir Oscar Hobson, City Editor. They have now been reprinted in booklet form, and copies are available, price 1*s.* 6*d.*, from the *News Chronicle* Book Department, 12/22 Bouverie Street, London, E.C.4. The articles survey the British steel industry and its position in the country's economy, and relate it to world markets; there is also an excellent short survey of new processes by Dr. T. P. Colclough.

New Furnace Company

Negotiations between Birlec Ltd. and Efco Ltd. for the formation of a new company, to take over certain sections of their business, have reached an advanced stage. The new company, to be known as Birlec–Efco (Melting) Ltd., will be responsible for the design and supply of all types of electric melting furnaces for the metal industries, together with smelting furnaces and induction heating equipment. The two parent companies will continue independently to manufacture their respective heat-treatment furnaces and other plant.

Institute of Welding

The Autumn Meeting of The Institute of Welding will take place at the Offices of the Institute, 54 Princes Gate, London, S.W.7, from 5th to 7th November. The Annual Dinner is being held at the Park Lane Hotel on the evening of Wednesday, 5th November; the Principal Guest will be the Rt. Hon. Aubrey Jones, P.C., M.P., Minister of Supply.

Opening of New Forge

The Rt. Hon. Lord Piercy, C.B.E., Chairman of the Industrial and Commercial Finance Corporation Ltd., performed the opening ceremony for the new light forge of Walter Somers Ltd., Halesowen, on 15th October.

Corrigenda

Owing to a printer's error, the second equation in the first paragraph on p. 315 of the paper by Dr. J. D. Hobson entitled "**The Diffusion of Hydrogen in Steel at Temperatures of − 78° to 200° C**" (*J.I.S.I.*, 1958, Aug., vol. 189) was incorrect; it should have been:

$$D = 1{\cdot}5 \times 10^{-3}\, e^{-6000/T} \text{ for } \gamma\text{-iron}$$

In the discussion on electrical maintenance in iron and steelworks (*J.I.S.I.*, 1958, Aug., vol. 189), there is an error in Mr. R. C. Stott's contribution. On p. 73, left-hand column, in the last paragraph of his contribution, line 3, there is a reference to "solid bearings or roller bearings"; this should read "sealed ball or roller bearings."

DIARY

3rd Nov.—Cleveland Institution of Engineers—President's Address, by T. H. Stayman—Cleveland Scientific and Technical Institution, Middlesbrough, 6.30 p.m.

4th Nov.—Sheffield Metallurgical Association, Sheffield Society of Engineers and Metallurgists, and The Iron and Steel Institute—"Furnace Design and Steelmaking Techniques," by C. Holden—B.I.S.R.A. Laboratories, Hoyle Street, Sheffield, 7 p.m.

5th–7th Nov.—Institute of Welding—Autumn Meeting—54 Princes Gate, London, S.W.7.

5th Nov.—Institute of Welding—Annual Dinner—Park Lane Hotel, London. W.1.

5th Nov.—Manchester Metallurgical Society—"Temperature Measurement," by J. A. Hall—Central Library, Manchester, 6.30 p.m.

6th Nov.—East Midlands Metallurgical Society—"Titanium—A Broad Survey," by P. L. Teed—College of Art, Green Lane, Derby, 7.30 p.m.

6th Nov.—Leeds Metallurgical Society—"Aluminium Bronzes," by C. V. Wilson—Chemistry Wing, University of Leeds, 7.15 p.m.

6th–7th Nov.—Verein Deutscher Eisenhüttenleute—Annual Meeting—Düsseldorf, Germany.

10th Nov.—Lincolnshire Iron and Steel Institute—"The Sendzimir Planetary Hot-Rolling Mill," by R. Habershon and H. W. Ward—North Lindsey Technical College, Scunthorpe, 7.30 p.m.

11th Nov.—Society of Chemical Industry (Chemical Engineering Group)—"The Integrated Iron and Steelworks," by W. B. Wright and W. Elland—14 Belgrave Square, London, S.W.1, 6 p.m.

12th Nov.—Institution of Engineering Inspection—"Metal Finishing," by A. W. Walbank—Birmingham Exchange and Engineering Centre, 6.30 p.m.

12th Nov.—Liverpool Metallurgical Society (Joint Meeting with the Institute of Welding, Liverpool Branch)—"The Gas Cutting of Metal Plates," by Dr. L. C. Bannister—Picton Library, Liverpool, 7 p.m.

12th Nov.—North Wales Metallurgical Society—"Gases in Iron and Steel," by C. E. A. Shanahan—Flintshire Technical College, Connah's Quay, 7 p.m.

12th Nov.—Sheffield Society of Engineers and Metallurgists—"The World's Largest Walking Dragline," by C. McL. Cameron—Sheffield University, 7.30 p.m.

13th Nov.—British Nuclear Energy Conference—"Electrification of U.K.A.E.A. Industrial Group Factories," by J. W. Binns and J. G. Outram—Institution of Electrical Engineers, Savoy Place, London, W.C.2, 5.30 p.m.

13th Nov.—Staffordshire Iron and Steel Institute—Presidential Address, by W. H. B. Wesson—Station Hotel, Dudley, 7.30 p.m.

16th–23rd Nov.—International Conference on Scientific Information—Washington, D.C., U.S.A.

17th Nov.—Sheffield Society of Engineers and Metallurgists—"Progress in Rod and Bar Rolling," by R. Stewartson—Sheffield University, 7.30 p.m.

18th Nov.—Newport and District Metallurgical Society—"A Glimpse at the Future of the Steel Industry," by Dr. C. A. Edwards, F.R.S.—Whitehead Institute, Newport, 7 p.m.

19th Nov.—Manchester Metallurgical Society—"Vacuum Melting," by H. C. Child—Central Library, Manchester, 6.30 p.m.

21st Nov.—Cleveland Institution of Engineers—Dinner—Teesdale Hall, Thornaby-on-Tees, 7 p.m.

26th Nov.—British Nuclear Energy Conference—"Recent Advances in Nuclear Engineering" (*Sir Charles Parsons Memorial Lecture*), by Sir John Cockcroft, O.M.—Institution of Marine Engineers, 76 Mark Lane, London, E.C.3, 5.30 p.m.

1st Dec.—Cleveland Institution of Engineers—"Basic Slag," by J. H. Thompson—Cleveland Scientific and Technical Institution, Middlesbrough, 6.30 p.m.

2nd Dec.—East Midlands Metallurgical Society—"Refining Nickel," by H. C. Castell—Electricity Showrooms, Smithy Row, Nottingham, 7.30 p.m.

3rd Dec.—Manchester Metallurgical Society—"Metallurgical Control in Mass Production," by J. G. James—Central Library, Manchester, 6.30 p.m.

4th Dec.—British Nuclear Energy Conference—"Thermonuclear Processes," by D. W. Fry—Institution of Electrical Engineers, Savoy Place, London, W.C.2, 5.30 p.m.

4th Dec.—Leeds Metallurgical Society—"Thermodynamics in Metallurgy," by Dr. J. Lumsden—Chemistry Wing, University of Leeds, 7.15 p.m.

5th Dec.—British Nuclear Energy Conference—"An Experimental Investigation into the Stress Distribution in a Band-reinforced Pressure Vessel," by K. G. Mantle, N. Marshall, and P. J. Palmer—Institution of Mechanical Engineers, 1 Birdcage Walk, London, S.W.1, 6 p.m.

6th Dec.—Liverpool Metallurgical Society—First Annual Dinner.

8th Dec.—Lincolnshire Iron and Steel Institute—"Engineering Aspects of a Modern Open-hearth Melting Shop," by J. A. Kilby—North Lindsey Technical College, Scunthorpe, 7.30 p.m.

9th Dec.—British Nuclear Energy Conference—"The Use of Radioactive Isotopes for the Investigation of Sewage Treatment Plant," by S. G. Burgess and A. F. Green—Institution of Civil Engineers, Great George Street, London, S.W.1, 5.30 p.m.

9th Dec.—Newport and District Metallurgical Society—"Oxygen in Steelmaking," by D. J. O. Brandt—Whitehead Institute, Newport, 7 p.m.

9th Dec.—The Iron and Steel Institute—11*th Hatfield Memorial Lecture*: "The Mechanical Properties of the Ferrite Crystal," by Dr. N. P. Allen, F.R.S.—Institution of Naval Architects, 10 Upper Belgrave Street, London S.W.1, 6.00 p.m.

10th Dec.—North Wales Metallurgical Society—"Metallurgy, Commerce, and Zinc Surfaces," by Dr. W. H. F. Tickle—Flintshire Technical College, Connah's Quay, 7 p.m.

10th Dec.—The Iron and Steel Institute—Autumn General Meeting—4 Grosvenor Gardens, London, S.W.1, 10.30 a.m.

11th Dec.—The Iron and Steel Institute (Iron and Steel Engineers Group)—Meeting on Vacuum Processes—4 Grosvenor Gardens, London, S.W.1.

11th Dec.—Liverpool Metallurgical Society—"The Metallurgical Applications of High-resolution Electron Microscopy," by Dr. J. Nutting—Department of Metallurgy, University of Liverpool, 7 p.m.

11th Dec.—North-East Metallurgical Society—"Bearing Metals from the User's Viewpoint," by R. J. Brown—Cleveland Scientific and Technical Institution, Middlesbrough, 7.15 p.m

BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on p. 208 of the October, 1958, issue of the Journal:

781. Šícha, M.: "Potentio-coulometric Determination of Small Amounts of Carbon in Steel and in Carbides." *Hutnicke Listy*, 1955, vol. 10, (9), pp. 535–542. (£6)

830. Lipovetskii, A., *et al.*: "Experience in the Adoption of Mechanized Coolers in Section Rolling Mills." *Stal*, 1956, (9), pp. 798–802. (£3 10*s.* 0*d.*)

864. Gruzinov, V. K.: "On the Movement of the Charge Column in the Blast-furnace." *Stal*, 1956, Sept., pp. 771–773. (£2)

884. Makelt, H.: "Two-roll Straightening and Multi-roller Straightening Machines for Round Bars, Tubes and Sections." *Werkstattstechnik u. Maschinenbau*, 1956, vol. 46, pp. 603–611. (£6)

887. Kuhn-Weiss, F.: "The Lubricating Properties and Industrial Application of Pure Molybdenum Disulphide." *Metalen*, 1957, vol. 12, Apr. 30th, pp. 148–153; May 15th, pp. 169–174. (£7 5*s.* 0*d.*)

891. Hroch, R., and A. Koval: "Casting of Steel Rolls for Blooming Mills." *Slevárenství* (*Prague*), 1957, vol. 5, June, pp. 161–168. (£5)

916. Part I. Boutonnet, P., *et al.*: "Basic Bessemer Steel Blown with a Mixture of Oxygen and Carbon Dioxide." Introduction and Part I: "Production of Steel with Low Nitrogen Content." *Rev. Mét.*, 1956, Aug., pp. 575–583. (£4 10*s.* 0*d.*)

949. Born, K.: "The Extraction of Sulphide Inclusions and Relation between the Properties of the Steel and the Inclusions." *Arch. Eisenhüttenwesen*, 1958, March, pp. 179–187. (£7)

963. Suchkov, I. A., and V. V. Burtsev: "Automatic Control of Gas Distribution in the Blast Furnace Shaft." *Stal*, 1958, Feb., pp. 110–113. (£3)

965. Branconi, V. von, and F. W. Griese: "Planning of the Handling of Bulk Materials in the Ironworks." *Stahl u. Eisen*, 1958, Apr. 17th, pp. 465–475. (£5 10*s.* 0*d.*)

966. Rosenbaum, K.: "Handling Plants and Storage Depots for Ore on its Way to the Harbour." *Stahl u. Eisen*, 1958, Apr. 17th, pp. 475–483. (£6 5*s.* 0*d.*)

975. Jaumotte, A.: "Two Multiple Gas Turbine Assemblies for a Belgian Ironworks." *Stahl u. Eisen*, 1958, May 15th, pp. 658–663. (£4 15*s.* 0*d.*)

977. Bogdandy, L. von: "An Examination of the Simultaneous Formation of Droplet Nuclei and the Growth of Droplets, illustrated by means of the Dust Formation in the Pneumatic Steel Refining Process." *Z. Elektrochemie*, 1957, vol. 61, (1), pp. 222–226. (£3 5*s.* 0*d.*)

984. Petrosyan, R.: "The Effect of the Wall Temperature of the Rear Surfaces of Boilers on Corrosion by Flue Gases." *Teploenergetika*, 1957, (12), pp. 16–20. (£3 15*s.* 0*d.*)

993. Bogdandy, L. von, and H.-D. Pantke: "Use of the Dust Produced during a Converter Blow as an Indicator of the Refining Process when Blowing with Oxygen-enriched Blast." *Stahl u. Eisen*, 1958, June 12th, pp. 792–798. (£5 5*s.* 0*d.*)

994. Höller, P., and H. Smit: "Optical Method of Determining Temperature Behaviour during Refining in a Converter." *Stahl u. Eisen*, 1958, June 12th, pp. 799–804. (£4 10*s.* 0*d.*)

998. Pich, R., and H. Haupt: "Determination of Measurements of Superheater Tubes with reference to Economic Considerations." *Energie*, 1957, Aug., pp. 287–293. (£4)

999. Orlov, S. I.: "Reducing the Size of the Crop Ends of Tubes when Rolling on Continuous Mills." *Stal*, 1958, Apr., pp. 335–339. (£4 5*s.* 0*d.*)

1002, 1004, and 1008 *Papers presented at the International Transport Conference, June* 1958, *Liège.*

1002. Mieth, H.: "A Comparison of the Performance, Lay-out and Cost of Different Works Railway Systems of German Foundries Supplemented by Mathematical Statistics." (£5)

1004. Forterre, R.: "Technical and Economic Aspects of Track Maintenance in a Steelworks." (£3 15*s.* 0*d.*)

1008. Hubin, L.: "Handling of Lime and Iron Scrap at Cockerill-Ougrée." (£2 15*s.* 0*d.*)

1019. Bungardt, K., *et al.*: "Investigations on the Structure of the System Iron–Chromium–Carbon." *Arch. Eisenhüttenwesen*, 1958, Mar., pp. 193–203. (£5 10*s.* 0*d.*)

1021. Assmann, H., and H. Buchhausen: "Materials used, Heat Treatment and Outputs of Thread Rolling Tools for the Manufacture of Screws." *Werkstatt u. Betrieb*, 1949, Nov., pp. 390–396; Dec., pp. 435–440. (£7 5*s.* 0*d.*)

1034. Bühler, H.: "Development of High-strength Steel Castings of Large Wall-thickness." *Giesserei*, 1958, Feb., pp. 87–92. (£3 5*s.* 0*d.*)

1035. Anon.: "The Reheating of Ingot-soaking Pits." *Tech. Moderne*, 1957, July, pp. 78–81. (£3 15*s.* 0*d.*)

1037. Borisov, S. I.: "Production of Alloy Steel Light-pressure Boiler Tubes." *Stal*, 1958, May, pp. 442–446. (£4 5*s.* 0*d.*)

1038. Möller, H., and H. Stäblein: "Rolling Textures of Transformer Steels." *Arch. Eisenhüttenwesen*, 1958, June, pp. 377–390. (£8)

1039. Gubkin, S. N., and I. A. Tsvaigel: "Deformation of Bronzes, subjected to Pressure." *Izvest. Akad. Nauk, Otdel. Tekh. Nauk*, 1954, Jan., pp. 128–137. (£2 15*s.* 0*d.*)

1042. Warth, H. von der, *et al.*: "The New Open-hearth Steel Plant No. III of the Hoesch-Westfalenhütte A.G., Dortmund." *Stahl u. Eisen*, 1958, Jan. 23rd, pp. 79–87. (£5 15*s.* 0*d.*)

1043. Geissler, K., and E. J. Kohlmeyer: "Contribution to the System Ferrous Oxide, Ferrous Sulphide, Manganous Oxide, Manganous Sulphide." *Arch. Eisenhüttenwesen*, 1958, Jan., pp. 57–63. (£6 15*s.* 0*d.*)

1044. Borodin, V. P., *et al.*: "Four-period System of Integrated Automatic Control for a Fuel Oil Fired Open-hearth Furnace." *Stal*, 1958, Feb., pp. 114–120. (£5 5*s.* 0*d.*)

1046. Zahorka, H.: "Progress in the Basic Lining of Glass Melting Tanks." *Radex Rundschau*, 1958, (3/4), pp. 154–172. (£4)

1047. Wesemann, F.: "Water Economy in Iron and Steel Works Undertakings." Lecture to the Main Session, Jernkontorets Tekniska Discussionmote, on 31st May, 1958. (£5)

1057. Rödl, W.: "Lining of Hot Metals Mixer." *Radex Rundschau*, 1958, (3/4), pp. 187–194. (£3)

1058. Fellcht, K., and R. Barz: "Metallurgical Control of Slag Inclusions and Carbide Segregations in Roller Bearing Steel." *Neue Hütte*, 1958, June, pp. 341–350. (£6 10*s.* 0*d.*)

1059. Witthof, J.: "Carbide Tools for Shaping and Cutting." *Draht*, 1957, (11), pp. 465–470. (£5)

1075. Nagami, K., *et al.*: "On the Distribution of Oxygen in Molten Steel during Refining." *Sumi. Met.*, 1958, vol. 10, Jan., pp. 1–4. (£1 5*s.* 0*d.*)

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects

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*Abstracts, the reference to which is marked with an asterisk are, or are to be made available as translations.

MINERAL RESOURCES

Geology and Magnetite Deposits of Dover District, Morris Co., New Jersey. P. K. Sims. (*U.S. Geol. Survey Professional Paper*, 287, 1958, pp. 162).

Mesabi to the North. (*Min. Eng.*, 1958, **10**, June, 678–679). A note on the Mt. Wright–Wabush Lake area iron deposits in Canada.

Iron Ore in India. G. H. S. V. Prasada Rao and Y. G. K. Murty. (*Iron and Steel Review*, 1957, **1**, Nov., 49–53). Of the world's total reserves of 85,000 × 10^6 t, India has 21,000 × 10^6 t. Details of various deposits are given and the present position of the Indian iron ore industry is reviewed.—A. G.

Chromite Deposits of India. M. N. Deekshitulu. (*Iron and Steel Review*, 1957, **1**, Nov., 59–63). The occurrence of chromite ores in India is reviewed and production, processing and future prospects are surveyed.

Manganese Industry in India. M. R. Subramanyam. (*Iron and Steel Review*, 1957, **1**, Nov., 55–58). The occurrence, characteristics, mining methods, reserves and exports of Indian Mn are discussed. The vital need for developing beneficiation processes is stressed.

Iron Ore Deposits—Koppio Area [South Australia]. R. K. Johns, (*South Australia Mining Review*, 1957, (103), 79–81).

ORES—MINING AND TREATMENT

The Future Supply of Iron Ore for Europe. A. Bureau. (*Soc. Ing. Civils Bul.*, 1958, (9), 188–193). In 1965 ore requirements for Europe including Gt. Britain are expected to be 70–77 × 10^6 tonnes. Increased supplies from the American continent and existing African mines should meet this demand. Thereafter, the levelling off of production in European mines will necessitate the development of new African mines.—A. G.

The Measurement of Deformations, Dilatations and Subsidences of the Surface in the Lorraine Iron Mines. E. Tincelin and P. Sinou. (*Rev. Ind. Min.*, 1958, **40**, Feb., 117–133).

Canadian Iron Ore. P. E. Cavanagh. (*Min. Eng.*, 1958, **10**, June, 680–682). Estimates of future production, consumption and exports for 1980.

Industrial Minerals Used in California's Iron and Steel Industry. K. W. Mote. (*Min. Eng.*, 1958, **10**, July, 765–767). An annotated list excluding ore, scrap, coke, Mn ore and ferro-alloys.

The Design of Bulk Carrying Vessels for the Australian Coastal Trade. D. A. Wittwer. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 6–10). Iron ore carriers are described.

The Operational Value of Iron Ore. P. M. Fourt. (*Ann. Mines*, 1958, Feb., 93–111). Calculations are shown for the choice between upgrading local ores or importing ores from overseas, and evaluated for the special case of Lorraine.

Blasting of Large Drill Holes in the Magnesite Deposits of the Millstätter Alps. H. Riedler. (*Radex Rund.*, 1958, (3/4), 113–119).

Dragonby Mine. (*Mine Quarry Eng.*, 1958, **24** Jan., 2–8; Feb. 46–51). This underground mine in the Frodingham Ironstone, near Scunthorpe, has been designed for an annual production of 10^6 t of ore, but is still in the development stage. It is worked by trackless mining methods, use being made of electric rotary drills, loaders, shuttle cars and a conveyor system that links the mining areas directly with rail wagon-loading facilities on the surface. A map, and many illustrations are given.

The Societe des Mines de Saizerais. (*Usine Nouv.*, 1958, **14**, July 31, 73–75). The history and present methods of operation of the iron mines at Marbache are outlined.

Quarrying Limestone at Cornelly, S. Wales. (*Iron Coal Trades Rev.*, 1958, **177**, Aug. 8, 336). A brief account of the Steel Co. of Wales operations.

Distribution Curves for Sink-and-Float Separation of Iron Ores. R. G. Wuerker. (*Min. Eng.*, 1958, **10**, July, 788–791). Laboratory tests are described and the method given and explained for obtaining theoretical optimum beneficiation. The distribution curves are shown to be the reverse of the washability curves for coal, (float-and-sink as compared with sink-and-float).

Iron Recovery Demands Engineering Compromises. C. H. Chilton. (*Chem. Eng.*, 1958, **65**, June 16, 132–135). Recovery from Ni ores is outlined.

Pressure Leaching of a Manganese Ore. Part I. Kinetic Aspects. Part II. Leaching Aspects. R. J. Cornelius and J. T. Woodcock. (*Proc. Australasian I.M.M.*, 1958, Mar., 65–107, 109–133). Aqueous suspensions of pyrite and Mn ore react with O_2 to dissolve Mn in the $FeSO_4$–H_2SO_4 produced. Optimum conditions are found.

Automation of Ore-Preparation Plants. G. Gerstenberger. (*Neue Hütte*, 1958, **3**, July, 389–398). The impact of automation on the progress of socialisation is emphasised. In the preparation of ore, however, automation will be of limited value as there are too many variables involved. The conditions under which the application of automatic equipment is justified are considered and summarised. A number of devices and apparatus for use in ore-preparation plant made in Russia and Eastern Germany is described. Only existing equipment is considered with regard to application in ore preparation.—T. G.

Reduction in Coke Consumption and Increase in Iron Production Effected by Ore Preparation. R. Mintrop. (*Stahl Eisen*, 1958, **78**, May 15, 633–646). The author compares German sintering practice with that employed in other countries. The total heat consumption per ton of iron produced cannot be reduced substantially, but high-grade coke can be saved by extensive sintering for which low-grade fuels are sufficient. The problems involved are interestingly elucidated in the discussion.—T. G.

Use of Mixed Firing in Sinter Plants. H. Rausch and K. Meyer. (*Stahl Eisen*, 1958, **78**, May 1, 600–606). The term "mixed firing" is used for a process in which part of the coke for sintering is replaced by hot gases produced by the combustion of liquid or gaseous fuels in a chamber above the sinter belt. Savings in coke of up to 40% were obtained, and total savings in calorific value amounted to 10–15%. The effect of temperature of the gases and the time of their action on sintering are discussed. The quality of the sinter is higher than that obtained in conventional processes. The throughput, however, is 5–10% lower.—T. G.

Second Sinter Plant for Iscor. (*Engineer Foundryman*, 1958, **23**, June, 35–38). An illustrated account of the installation at Pretoria.

The Use of Pyrites Ash in the Production of Fluxed Sinter of High Basicity. L. Ya. Levin, I. M. Yakubtsiner, V. M. Sholeninov and G. F. Grigor'evykh. (*Metallurg*, 1958,

(6), 5–10). Due to the shortage of ore concentrate, at the Cherepovets plant, pyrites ash was added to the sinter mix in the proportion of 25% of the ore addition. Optimum conditions for maximum sulphur removal and the production of high strength sinter are discussed.—R. S.

Production of Sinter by Gaseous Sintering. A. K. Rudkov. (*Metallurg*, 1958, (4), 3–5). Consumption of solid fuel is halved with the process and gas consumption is increased so that the economics depend upon local conditions, such as relative costs of blast furnace gas and coke fines and availability of gas. The cost of sinter is slightly higher at Dzerzhinsk, where the process is used, but this is compensated for by increased reducibility and saving of coke in the blast furnace.

Effect of Ignition on the Pore System of Iron Ore Lumps. A. E. El-Mehairy. (*Mining Engineering, Cairo Univ.*, 1957, reprint. 297–306). [In English]. A haematite was examined in the dry and ignited states by total porosity measurement, gas permeability and computation of average pore size from air flow measurements using the viscous slip phenomenon. Extent of interconnection of the pores is increased by an amount related to loss on ignition.

Microstructures of Iron Ore Sinter. H. C. Coe and D. H. Skinner. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 2–5). Metallographic methods were used and structures are shown and related to composition.

Iron Ore Reducibility Tests. (*Min. Eng.*, 1958, **10**, June, 682). The Ontario Research Foundation and Russian methods are outlined and a new test is suggested using a reducing gas such as CO.

FUEL—PREPARATION, PROPERTIES AND USES

Energy Distribution in an Integrated Iron and Steel Works. K. Rieder. (*Stahl Eisen*, 1958, **78**, May 15, 669–671). The author analyses the main energy carriers of the Hütte Linz of the Vereinigten Österreichischen Eisen- und Stahlwerke, such as coke-oven gas, electric power and flue gas. 2000×10^6 k. cal. are converted to a different form of energy every hour. The analysis is illustrated by diagrams and flow sheets.—T. G.

Power Economy of the West-German Iron- and Steel-Industry During the Years 1950–57. H. Wübbenhorst. (*Stahl Eisen*, 1958, **78**, June 12, 777–784). The total heat consumption per ton of steel produced has been reduced in the period from 1950 to 1957 by 15%. This is mainly due to increased production, change in the burden, technical improvements, and use of the most suitable fuels for each process.—T. G.

The Petrography of Coal and Its Relationship to Fuel Technology. I. Origins of Coal Petrography. V. H. Rodriguez. (*Inst. Nac. Carbon Bol. Inf.*, 1957, **6**, May–June, 115–118). Historical. **II. Instruments for Investigation.** (119–135). A detailed review of microscopical techniques using reflected light, and U.V., phase, electron microscopy and radiographic methods. **III. Techniques of Microscopical Investigation.** (Nov.-Dec., 274–279). Methods of specimen preparation other than thin section preparation are described.

Research in the Development of a Metallurgical Fuel from Brown Coal. G. L. Kennedy and D. G. Evans. (*J. Australian Inst. Met.*, 1958, **3**, May, 61–75). The application of German briquetting and carbonization techniques has enabled brown coal briquettes to be made into a hard lump char which has been successfully used in a cold-blast cupola.

Determination of the Optimal Ash Content of Blast Furnace Coke and the Optimal Water Content of Coking Coal. K. Lemke. (*Glückauf*, 1958, **94**, June 21, 817–824). Costs for reducing the ash content of coking coal below 10% and the total water content below the same amount were calculated for three coals, and the results discussed.—L. D. H.

Relationships between [Coal] Preparation, Coke Oven Plant and Blast Furnace. W. Reerink and K. Lemke. (*Glückauf*, 1958, **94**, June 21, 832–842). This report is the result of work carried out in several countries, and is concerned with the possibilities of suitable preparation of the coal and design and operation of the coking plant in producing a uniform blast furnace coke of good quality. Economic aspects are discussed in detail. (26 refs.)—L. D. H.

Repositioning a Coke Ovens Stack. H. Hughes. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 21–24). Moving the stack at No. 1 Battery, Newcastle (N.S.W.) steelworks is described. The steel stack weighing 208 t, with 64 t brickwork, was moved 50 ft.

Modern Developments in Steelworks Gas Producer Practice. J. C. Martin. (*J. B'ham Met. Soc.*, 1958, **38**, June, 76–81).

Contribution to the Study of the Composition of Blast-Furnace Gas. Application to the Calculation of Coke Consumption. A. Decker and L. Bousmanne. (*Rev. Univ. Mines*, 1958, **14**, Mar., 80–105). General formulae are derived for all components of gases, with numerous curves and tables.

Desulphurization of Coke-Oven Gas at Appleby-Frodingham. L. Reeve. (*J. Inst. Fuel*, 1958, **31**, July, 319–324). Hot Fe_2O_3 in static and fluidized beds is used.

AIR POLLUTION AND SMOKE

The Prevention of Atmospheric Pollution. J. W. H. Holmes. (*Guilds Engineer*, 1958, (9), 73–83). A review of dust removal with notes on SO_2.

Evaluation of Air-Borne Particulates in Atmospheric Pollution Studies. M. Katz, H. P. Sanderson and M. B. Ferguson. (*Anal. Chem.*, 1958, **30**, July, 1172–1180).

Notes on the Removal of Particulate Matter and Vapour Phase Sulphur Compounds from Gases. R. J. S. Thompson and W. F. Cosby. (*Gas World*, 1958, **148**, Aug. 2, *Coking*, 23–26, 28, 30).

Measurement of Traces of Gases in the Atmosphere of Iron Works. F. Bangert. (*Stahl Eisen*, 1958, **78**, May 29, 743–747). The author describes modern gas detection equipment and its use for hygienic and operational reasons, mainly for the detection of CO. Automatic, continuous measurement and alarm equipment is also dealt with.—T. G.

TEMPERATURE MEASUREMENT AND CONTROL

Methods of Testing Thermocouples and Thermocouple Materials. W. F. Roeser and S. T. Lonberger. (*N.B.S. Circular* 590, 1958, pp. 21). A review covering noble and base-metal couples.

Temperature of the Flame of a Single Drop of Liquid Fuel. M. A. Gurevich and V. B. Shteinberg. (*Zhur. Tekhn. Fiz.*, 1958, **28**, Feb., 394–401).

An Optical Pyrometer with Emissivity Compensation. J. Euler. (*Elektrowärme*, 1958, **16**, June, 194–199). A method of obtaining true temperature from a disappearing filament pyrometer is described. An auxiliary source is projected on to the hot body through an aspherical mirror and the aperture is adjusted for distance, emissivity and quality of mirror and lenses. The instrument is most suitable for short distances and as a micropyrometer. It reads to a few ° C.

The Design of an Immersion Thermocouple and Its Use in the Open-Hearth Furnace. H. Löscher. (*Neue Hütte*, 1958, **3**, July, 416–425). The design of an immersion thermocouple using Pt–6%Rh/Pt–30%Rh is described and results obtained in O.H. furnaces are reported. No condensation products from the thermocouple were observed in the quartz sheath. (28 references).—T. G.

A Study of "Freezing" of Pyrometric Cones. Chen Hsen-ku. (*Acta. Met. Sin.*, 1957, **2**, (1), 99–106). [In Chinese]. Freezing (not squatting at the softening temp.) is caused by the formation of crystals from molten phases in the cone when the furnace temp. is unsteady or remains for a long time below the softening temp. In SK7–10 cones these are cristobalite and in SK 11–16 may also contain mullite. They can be avoided by reducing the proportions of Na_2O or K_2O, respectively; CaO or MgO cannot be substituted for the alkali oxides.—K. E. J.

REFRACTORY MATERIALS

Refractory Products. Y. Letort. (*Usine Nouv.*, 1958, Spring, 119–124). A brief account is given of the French refractory industry, technical progress and method of studying refractory products. The various types of refractory are listed and their properties and applications are discussed.—A. G.

What's New in Refractories. (*Steel*, 1958, **142**, Mar. 31, 89–90). Recent trends in steelworks refractories are described. Quality control during manufacture has improved product quality. There is a continued trend towards the all-basic O.H., and after an initial success, more roofs of fused bricks are being tried; a carbon sub-hearth has shown advantages. High-capacity O_2 converters are using tar bonded dolomite brick. The use of carbon blast-furnace linings is spreading for hearth-walls and bosh; standard brick sizes are sometimes replacing large blocks.—D. L. C. F.

Research on Refractories for Open Hearth Furnaces. (*Refract. J.*, 1958, **34**, June, 278–279). From the B.I.S.R.A. 1957 annual report.

Electric Furnace Refractories. E. C. Wright (*Met. Prog.*, 1958, **73**, May, 87–88). The proceedings of the Electric Furnace Steel Conference at Pittsburgh are summarized. Cast and rammed sections in O.H. roofs, the use of Canadian magnesite tar-bonded basic and ladle refractories are discussed.

High Alumina Refractories: Use for Steel Pouring in Russia. K. Shaw. (*Iron Steel*, 1958, **31**, May, 184–185). A technique has been developed in Russia for producing mullite-corundum refractories having high-temperature stability and a constant volume at high temperatures. Details are given of trials of the refractories as ladle linings and stoppers.—G. F.

Control of Factors Affecting Reproducibility of Mechanical Properties of Refractory Semi-dry Press Specimens. L. E. Mong and D. M. Adelman. (*J. Amer. Ceram. Soc.*, 1958, **41**, July, 267–272).

Refractory Materials in India. G. G. K. Sastri. (*Iron and Steel Review*, 1957, **1**, Nov., 65–69). Raw materials and prospects for the refractories industry are surveyed.

Zirconia–Crystobalite: A New Acid Lining Material for the Steel Industry. E. Steinhoff. (*Stahl Eisen*, 1958, **78**, June 26, 893–899). Zirconia–crystobalite bricks have excellent thermal-shock resistance; they do not grow at high temperatures but show a slight shrinking, for this an allowance must be made in the construction of roofs and covers. Chemical analysis showed that in service the surface of the bricks becomes enriched in zirconia thus increasing the viscosity of the surface layer and preventing drop formation. The wear mechanism of the zirconia–crystobalite bricks is explained.—T. G.

Hydration of Basic Refractories. G. R. Eusner and J. R. Bachman. (*Amer. Ceram. Soc. Bull.*, 1958, May, 213–219). Autoclave tests for testing resistance to hydration are described on various bricks and autoclave, reflux-boiling and a new burn-in test on ramming mixes. A possible mechanism of disruption of the latter is disclosed. Atmospheric stabilities are considered.

Research on Super Duty Insulating Fire Brick. I. E. S. Nagai, G. Yamaguchi, Z. Ota, F. Tanemura and M. Tsukino. (*Ann. Rep. Eng. Res. Inst. Tokyo Univ.*, 1957, **16**, Sept., 89–92). [In Japanese]. Optimum conditions are established for manufacture, by sintering, of a refractory formed from diaspore, fireclay, aluminium powder, dilute aq. NaOH and water-glass soln. The refractory had a bulk density of 0·7–0·9, and good insulating qualities.—K. E. J.

Determination of Phases in Silica Bricks. Morand and Kiehl. (*Silicates Ind.*, 1958, May, 248–252). A method is described for the rapid quantitative estimation of the different forms of silica in silica bricks. Use is made of X-ray diffraction with a Geiger counter and alumina is added as internal standard.—A. G.

Research on the Silicate Minerals in Dolomite Refractories. I. G. Yamaguchi and S. Komatsu. (*Ann. Rep. Eng. Res. Inst. Tokyo Univ.*, 1957, **16**, Sept., 86–88). [In Japanese]. The mineral constituents are reported for mixtures of CaO and SiO_2 in the molar ratios 1 : 1, 2 : 1 and 3 : 1, fired at 1500° C.—K. E. J.

The Effect of Various Factors on the Upper and Lower Transformation Temperatures of Cristobalite. Chang Shou-Ching, Ho Chung-Fan and Kiang Chien-Chia. (*Acta Chim. Sinica*, 1957, **23**, (6), 462–468). [In Chinese with German Summary]. The effects of particle size, temperature of formation, raw material and the effect of mineralising agents (Li_2CO_3, Na_2CO_3, Na_2WO_4, FeO, CaO in decreasing order of effectiveness) are investigated.

Subsolidus Relations between Mullite and Iron Oxide. W. E. Brownell. (*J. Amer. Ceram. Soc.*, 1958, **41**, June, 226–230). The study had special reference to the fired colour of clay products.

Some Microscopical Observations of the Behaviour of Chromite in Refractory Chrome Magnesite Brick. F. Trojer. (*Radex Rund.*, 1958, May, 123–130). A report on an investigation under the petrographic microscope of the behaviour of chromite grains in chromite refractories during use in industrial furnaces. The processes occurring are illustrated and discussed.

Determination of Mullite in Clay Refractory Products. Yeh Fu-han. (*Acta Met. Sin.*, 1956, **1**, (4), 401–405). [In Chinese].—K. E. J.

On the Stability of the Phase Fe_2O_3, Al_2O_3. A. Muan. (*Am. J. Sci.*, 1958, **256**, June, 413–422). The work was carried out with special reference to steel plant refractories.

Tests on Zirconia-Silica Bricks for Arc-Furnace Covers. A. Theis. (*Stahl Eisen*, 1958, **78**, June 26, 891–892). The bricks are made from zirconia and crystobalite and have the following physical data (values of conventional silica bricks in brackets): Softening point 1750° C (1700° C); porosity 23% (20%); sp. gr. 2·56 (1·9); thermal expansion 0·86 (1·4). The service life of the new bricks is about twice as long as that of silica bricks, parts of damaged covers can be repaired so that the service life of the covers could be extended to six times those of silica bricks.—T. G.

Stressing and Durability. A Symposium of the Most Recent Knowledge on the Use of Basic Furnace Refractories in High Temperature Furnaces. L. Hütter. (*Radex-Rund.*, 1958, May, 135–145). A general survey of recent work on the use of basic refractories in various types of high temperature furnace, as described in the literature in recent years. (24 refs.).

High Temperature Strength of Basic Refractories. E. I. Greaves and J. Mackenzie. (*Trans. Brit. Ceram. Soc.*, 1958, **57**, Apr., 187–198). Methods and results of torsion and tensile tests are described. Good correlation was found between high-temperature strength, creep and service in O.H. roofs. The torsion test in particular is suggested as a measure of one factor in brick stability for roof service.

An Apparatus for Determining the Thermal Conductivity of Refractories. D. M. Shakhtin. (*Zavods. Lab.*, 1956, **22**, (7), 869–871; translation 19*RL–T*/C67).

Emission of Silico-Aluminous Refractories in the Infra-Red Region. R. Alegre. (*Silicates Ind.*, 1958, May, 253–260). The diffuse reflection of these refractories has been studied at wavelengths from visible to 12 μ and at temperatures up to 1000° C. Around 3 μ, reflectivity is low leading to marked absorption of heat from certain flames such as town's gas.—A. G.

The Use of Statistical Methods in the Quality Control of a Works in the Refractories Industry. F. Kiessewetter. (*Radex Rund.*, 1958, May, 131–134). A short discussion of the use of mathematical statistics in sampling raw magnesite, the application of control charts and calculation of regression coefficients to the control of the manufacturing process and the evaluation of the results.

IRON AND STEEL, GENERAL

Developments in the Iron and Steel Industry during 1957. I. E. Madsen. (*Iron Steel Eng.*, 1958, **35**, Jan., 139–189). A comprehensive review is given of plant and process developments in the U.S. and elsewhere in the world, including a tabulated list of expansion plans announced by U.S. companies.—K. E. J.

The Siting of British Steelworks. (*Steel Rev.*, 1958, July, 26–35). A general review of factors, with maps showing locations of markets, labour, scrap, ores, coal and water.

Europe Expands Steel-Making Capacity. D. L. McBride. (*S.A.E.J.*, 1958, **66**, June, 62–63). A review.

The Czechoslovak Metallurgical Industry and its Perspectives. J. Hlavaček. (*Czech. Heavy Ind.*, 1958, (6), 7–11). Mainly iron and steel. Plants and projected constructions are noted with a few statistics and notes on products.

The Manufacture of Pig Iron and Steel in Czechoslovakia. (*Czech. Heavy Ind.*, 1958, (6), 12–20). A general illustrated review.

The Steel Industry of Mexico—its present and Immediate Future. P. G. Roldán. (*Iron Steel Eng.*, 1958, **35**, Jan., 80–84). A brief survey is made of the availability of raw materials and fuels. The present source of iron ore is Cerro del Mercado (average 60% Fe); proved deposits there and elsewhere amount to 330 × 10^6 t, and others remain to be explored. In 1956 Mexico produced 865,000 (short) t of ingots, and output should reach at least 2,000,000 t by 1965. Details are given of the plant at the six major steelworks.—K. E. J.

The Mexican Iron and Steel Industry. (*Mét. Constr. Mécan.*, 1958, **90**, May, 333–340). After a discussion of Mexican reserves of ore, coal and coke, details of current production of iron and steel are given. Future development will be based on the integration of Northern works and the creation of a new industry based on Pacific ore reserves.—A. G.

The Activities of the Verein Deutscher Eisenhüttenleute During 1957. (*Stahl Eisen*, 1958, **78**, May 1, 565–599). The report deals *inter alia* with: Blast-furnace practice, Steelmaking, Rolling-mill practice, Forging, Cold-rolling, Descaling, Wire drawing, Galvanizing, Metallurgy and related fields, Powder metallurgy and History. In general, the report summarises papers and discussions published and to be published in *Stahl u. Eisen* and *Arch. Eisenhüttenwesen*.—T. G.

BLAST FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Examination of Blast Furnace Production with Radioactive Isotopes. V. A. Smolyak and V. N. Uzlyuk. (*Metallurg.*, 1958, (4), 7–9). The study of the movement of fine fractions of burden materials and measurement of level of smelting products in the hearth are described.—R. S.

The Development of Blast Furnace Technique in the U.S.S.R. Z. Punčochař. (*Hutnik (Prague)*, 1958, **8**, Apr., 115–117). The present state of blast furnace technique in the Soviet Union is briefly outlined. It is reported that blast furnace utilization has reached a level of 78–80%, and coke consumption per t pig-iron for steel-making has fallen to 655 kg. as a result of improved coke quality. Extensive use is made of self-fluxing sinter, and the total quantity of sinter amounted to 66% of the total furnace input of ore and sinter in 1956; scrap input was only 53 kg./t. Mention is made of refractories with very high properties, oxygen enrichment of the blast, and desulphurization by the insertion of pulverised lime into the ladle. Much space is devoted to a description of large new blast furnaces in operation or under construction in the U.S.S.R. with diagrams and dimension table. For the largest, to be completed this year (2286 m^3 capacity) 6–15% coke-oven gas will be added to the blast furnace gas, and the blast will be enriched with 25% oxygen; the blast furnace air will be humidified with 5·3% water vapour, and the stove will preheat the air to 1200° C. With a gas pressure of 1·5–1·8 atm. and a blast pressure of 3·2 atm., it is intended to raise the temperature below the bell to 1500° C. The degree of automation will be greatly increased.—M. T.

Ironmaking: Art or Science? H. A. Goldfein. (*Blast Furn. Steel Plant*, 1958, **46**, June, 611–614). In this lecture, the author discusses the numerous variables encountered in the day-to-day operation of a blast furnace. He stresses that the foreman must recognize the problems and understand how each fits into the total picture so that the end product may be quality iron.—T. G.

Pig Iron Production. (*Engineer Foundryman*, 1958, **23**, April, 34–36). The extensions to the blast furnace plant of the African Metals Corporation Ltd. are described. The capacity of the new plant will be 1000 t/day.

Use of Carbon Refractories in Blast Furnaces. P. I. Lokshin and G. P. Moshkina. (*Stal'*, 1956, (2), 108–114). The use of carbon refractories in the U.S.S.R. and abroad is reviewed. Linings of hearth walls and bottoms are most often satisfactory especially where bottoms are all carbon, cooled at sides and underneath. The British design is regarded as the best. Shaped graphitized blocks with high thermal conductivity might give the best results. A design of hearth bottom is shown and the best structure in the absence of shaped blocks is given.

Block Charging of Blast Furnaces. V. E. Levchenko. (*Metallurg.*, 1958, (5), 4–6). K. A. Bobkun (6–8). Block charging not only increases the drive of the furnace but lowers coke consumption by ensuring correct distribution of the gas streams transversely in the furnace. The effects in relation to the amount of fines in sinter and ore are discussed. R. D. Kamenev and V. N. Fomenko (8–10). A proposed method of charging the new 2000 m^3 furnace using conveyor belts is discussed and disadvantages are pointed out. Methods for charging coke and sinter or coke, sinter and other components are suggested.

Recording the Operation of the Charging Gear of Blast Furnaces. E. I. Sametskii. (*Metallurg*, 1958, (6), 4–5). A self-recording device to register the angle of rotation of the distributor to within 1·5°, the lowering of the small and large bells, the number of ore and coke charges and the times of each operation is described.—R. S.

A Comparison of Pneumatic and Electric Bell-Operating Machinery. L. J. Kilboy, jun. (*Blast Furn. Steel Plant*, 1958, **46**, June, 606–610). Recently very few electric bell hoists have been installed. Since 1940 twice as many pneumatic hoists were installed as in the previous twenty years. Approximately 75% of U.S. blast furnaces have pneumatic bell hoists. About 5% employ electric hoists and the remaining 20% utilize various types of steam or hydraulic devices. Advantages and shortcomings of the pneumatic and electric types of hoists are enumerated and discussed.—T. G.

Shock Absorbing Device for Blast Furnace Bell Balance Beams. L. V. Brodetskii. (*Metallurg*, 1958 (6), 3–4). A shock-absorbing device for improving the elastic working of the balance beams of the small and large bells at Kuznetsk is described.—R. S.

Furnace is 'Quickfrozen.' (*Steel*, 1958, **142**, June 16, 106). An experimental blast furnace has been quenched using a prolonged blast of nitrogen. Preliminary sampling of the burden has shown (*a*) unaltered coke penetrates deep into the slag-metal layer and (*b*) raw materials remain stratified as charged until they reach the melting zone.—A. G.

Automatic Control of Blast-Furnace Operation. A. N. Makarow. (*Neue Hütte*, 1958, **3**, July, 404–408). The author reports on current practice of iron smelting in automatically controlled furnaces in the Soviet Union. In 1955 about 90% of the total pig iron produced came from automatically operated furnaces. The author explains by means of diagrams the principles of measuring and controlling the various reactions taking place in the blast furnace, and discusses automatic control of the charging operation, of the blast temperature and humidity and of

the heat balance. An ultrasonic instrument for measuring the height of the burden is also described.—T. G.

The Theory of Thermodynamic Equilibrium in the Blast Furnace and Its Application. P. Reichardt. (*Stahl Eisen*, 1958, **78**, May 29, 748–750). The author reviews and discusses a theory by J. Raick. A number of points that are not in agreement with the author's own views on this subject are stressed.

Blast Furnace Smelting with High MgO Slag. Chuang Cheng-woo. (*Acta Met. Sin.*, 1957, **2**, (3), 219–235). [In Chinese]. When smelting with slag containing 18·5% MgO, the furnace runs smoothly with stable hearth temp. The desulphurizing power is low; to raise it, the total basicity (CaO + MgO)/SiO_2 must be raised. The higher coke consumption with dolomite in the burden is caused by the 15% increase in CO_2, which increases the amount of direct reduction. Dolomite should be used in the sinter.—K. E. J.

An Investigation of Chemical Equilibria in the Blast-Furnace Hearth. I. The Manganese–Sulphur Equilibrium. Chou Yuan-hsi. (*Acta Met. Sin.*, 1956, **1**, (4), 325–335). [In Chinese]. Data from 122 tests show that the Mn–S equilibrium in the hearth is in agreement with theoretical predictions. The equilibrium constant is related to the activities of CaO, MnO and CaS. Some results are given for the effects of basicity and (S) on the equilibrium and for the temperature-dependence of the equilibrium constant. (13 references). **II. The Silicon–Sulphur, Silicon–Carbon and Silicon–Manganese Equilibria.** (1957, **2**, (1), 13–24). With the aid of recent data, it is shown that neither the Si–S nor the Si–Mn equilibrium is approached in the hearth. By using $\sqrt{[\%\ Si]}/\sqrt{(a_{SiO_2})}$ as an index of the degree of silicon reduction, it is found that an increase in slag basicity promotes the reduction. Thus the Si–C reaction may approach equilibrium if the basicity is high.

Dealing With Cyanides in Blast Furnace Top Gas. S. P. Koshelev. (*Metallurg*, 1958, (5), 11). When a blast furnace at Novo Tula was turned over to ferromanganese the cyanide content of the gas increased, affecting labour conditions near the furnace and poisoning the fish in the river. The content had risen to 1 g/m³ due to extension of the high temperature region of the furnace and to the effect of alkali metals in the ore and coke ash. Chiatursk ores contain 0·1–0·2% alkalis after washing (0·5–1·1% run of mine). With smooth operation cyanide was 0·072–0·234 mg/l. When top gas temperature was raised by 46° C cyanide rose 30% (with 1·29% Si). Increase of Si to 1·85% with constant top gas temperature increases cyanide to 0·082–0·165 mg/l. Top gas temperature should therefore be kept down. Two methods of treatment for waste water are given.—R. S.

High Top Pressure Above One Atmosphere. V. P. Onoprienko, B. N. Starshinov, P. G. Netrebko, D. S. Yaloboi and G. B. Rabinovich. (*Metallurg*, 1958, (4), 6). Tables are given of the operation of a Ukrainian blast furnace. The top pressure was raised successively from 0·46 atm. (March, 1956) to 0·71 in July, 1·05 in August and 1·13 in October. Iron composition is given and Mn ore consumption did not exceed 28 lb/t of iron.—R. S.

New Knowledge in Connection with the Production of Pig Iron in the Low Shaft Furnaces of the German Democratic Republic. G. von Struve and R. Ebert. (*Bergakademie*, 1958, **10**, Apr., 229–235, reprint). After outlining the development of the various types of low shaft furnace and their underlying principles, the main characteristics of the furnaces at Calbe are described, with particular reference to the distribution of the burden. It was found that Mn reduction was as complete as in the blast furnace, but the Mn yield is unsatisfactory due to heavy slag bulk and dust losses. Sulphur and phosphorus reduction are not as satisfactory, and indirect reduction is much less than that in the blast furnace. In certain circumstances oxygen-enriched blast has been used successfully.

Electric Smelting of Iron Ore. K. Sandvold, F. C. Collin and J. Gundessen. (*Chim. Ind.*, 1958, **79**, Feb., 144–149). Practice at Mo i Rana, and Tysland-Hole furnace operation or construction are outlined. Material preparation is stressed and the relative possibilities of the electric and coke-fired furnaces are considered.

Alternative Processes of Making Iron and Steel. A. G. Robiette. (*Iron Coal Trades Rev.*, 1958, **177**, July 25, 219–216). A review of the processes alternative to blast-furnace production, in operation and in the development stage.

Pre-Treatment of Molten Iron. Desiliconization and Dephosphorization Experiments. L. Cook. (*Iron Coal Trades Rev.*, 1958, **177**, July 11, 87–93). The use of the oxygen lance with simultaneous injection of lime or limestone, millscale, fluorspar and bauxite in various combinations is investigated. Complete quantitative data and analyses are tabulated.

The Lining of Iron Mixers. W. Rödl. (**Radex Rund.*, 1958, May, 187–194). A general discussion of various aspects of the selection of materials, design, construction and performance of iron mixers.

PROPERTIES, TREATMENT AND USE OF SLAGS

Thermodynamics of Liquid Blast-Furnace Type Slags. I. Evaluation of Activities of CaO and SiO_2. II. The Application of a New Basicity Ratio to Desulphurisation of Carbon-saturated Iron. Chou Yuan-hsi. (*Acta Met. Sin.*, 1956, **1**, (2), 127–141; 142–154). [In Chinese]. I. New data are presented for activities of CaO and SiO_2 in liquid CaO–SiO_2, CaO–SiO_2–Al_2O_3 and CaO–MgO–SiO_2–Al_2O_3 systems. Activities of CaO are plotted against (%CaO)/(%SiO_2). The curves lead to a new expression for basicity, in which CaO activities in all systems become approximately equal at the same basicity. Al_2O_3 shows amphoteric behaviour. (23 references). **II.** The new basicity ratio gives useful results when correlated with data for desulphurization power. (11 references). **III. Evaluation of the Activity Coefficient of MnO.** (*Acta Met. Sin.*, (3), 233–242). The activity coeff. of MnO in liquid B.-F. type slags is evaluated as a function of a new basicity ratio, with reference to pure solid MnO as the standard state. The results are discussed from the viewpoint of the ionic nature of slags. (13 references).—K. E. J.

Study on the Activity of Constituents in the Molten Slag. I. On the Activity of Silica in the Systems: CaO–SiO_2, CaO–SiO_2–Al_2O_3. K. Sanbongi and Y. Ohmori. (*Nippon Kinzoku*, 1957, **21**, Apr., 296–300). [In Japanese]. Studies in a concentration cell show a break corresponding with CaO.SiO_2 on the e.m.f.-composition curve for the CaO–SiO_2 system. The equation giving the activity coeff. of SiO_2 is deduced. Iso-activity lines in the CaO–SiO_2–Al_2O_3 system are discussed. It is assumed that the anionic species of SiO_2 in the system are SiO_4^{-4} and SiO_3^{2-}.—K. E. J.

A Study on the Diffusion of Calcium Ion in Lime–Silica–Alumina Slags by Using a Radioactive Tracer Technique. K. Niwa. (*Nippon Kinzoku*, 1957, **21**, Apr., 304–308). [In Japanese]. Using ^{45}Ca as tracer, it is shown that the activation coeff. of the diffusion of Ca^{++} is 30 ± 15 k cal in a slag containing 43% CaO, 39% SiO_2 and 18% Al_2O_3.

Equipment for the Study of Hot Slags. F. Adcock. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 11–12). A method for heating minute slag samples on the microscope stage by means of Pt wires is described.

DIRECT PROCESSES

The Direct Reduction of Iron Ore in the United States. H. W. McQuaid. (*Iron Steel Eng.*, 1958, **35**, Jan., 69–77). Present capital costs are disadvantageous to the B.F. and O.H. furnaces, and increased steel demand can be met from electric arc furnaces. In view of high scrap costs and erratic supplies, however, there is great interest in direct-reduction processes. A technical and economic survey is made of recent work on such processes, especially those based on hydrogen and natural gas as fuels.—K. E. J.

Rebuilding and First Results of the Krupp-Renn Plant at Salzgitter-Watenstedt. D. Fastje. (*Stahl Eisen*, 1958, **78**, June 12, 784–792). The Krupp-Renn plant demolished after the war was rebuilt in 1955–57, the first of the three kilns was put into operation in Oct., 1956, the last in March, 1957. Layout, modification of the plant compared with the old one, and results of the first few months of operation are described. In 1957, production amounted to about 140,000 t of puddled bars with an average iron content of 92·5%.—T. G.

The R-N Process for Direct Reduction of Iron Ores. (*Met. Prog.*, 1958, **73**, April, 73–77). A pilot plant jointly developed by the Republic Steel Corp. and the National Lead Co. at Birmingham, Ala., is described. Fine ore is reduced directly with coke breeze, anthracite duff, etc., to produce a briquetted concentrate containing 90–95% total Fe and 3–8% SiO_2 which is suitable for further processing in blast furnaces, cupolas or O.H. or electric furnaces.

Procedures for the Direct Reduction of Iron Ore and Melting Generally. Fifth Committee. General and Technical Studies. E. Aguilar. (*Second National Congress of the Iron and Steel Industry*, [Mexico], 1957, 101–109). The "H" hydrogen reduction process is described and costs are given.

PRODUCTION OF STEEL

Twenty Years of Steelmaking. R. C. Baker. (*Iron Steel*, 1958, **31**, May, 186–192). The author describes the changes made in steelmaking plant and procedure at John Baker & Bessemer Ltd. during the past 20 years. Particular reference is made to instrumentation and automatic control.—G. F.

New Developments in Steelmaking. J. S. Curphey. (*Refract. J.*, 1958, **34**, June, 259–272, discussion 273–275, 277). A review of recent processes with reference to furnace and other vessel construction and refractory linings.

Open-hearth Furnace Regenerators: A Survey of Theoretical and Experimental Work, with Recommendations for the Calculation of Preheat. B.I.S.R.A. Regenerator Group. (*J.I.S.I.*, 1958, **190**, Nov., 254–271.) [This issue.]

Mechanising the Preparation of Powder Materials in Electric Steelmaking. A. I. Sapko, V. P. Dobrov and V. D. Kucherov. (*Metallurg*, 1958 (6), 20–22). The authors outline some defects in methods of crushing ferro-silicon at electric steel plants and suggest some improvements.—R. S.

Research on the Desulphurising Properties of Ochre. I. G. Yamaguchi and H. Abe. (*Ann. Rep. Eng. Res. Inst. Tokyo Univ.*, 1957, **16**, Sept., 92–94). [In Japanese]. Chemical and thermal analyses show that samples are crystalline or amorphous Fe_2O_3.H_2O. Desulphurizing power is attributed to amorphous hydrated Fe_2O_3.—K. E. J.

The Vacuum-lifter Process of the Dortmund-Hörder Hüttenwesen AG. F. Harders. (*J.I.S.I.*, 1958, **190**, Nov., 306–311.) [This issue.]

Reducing the Consumption of Aluminium for the Deoxidation of Steel. A. M. Ofengenden and R. P. Nesterovich. (*Metallurg*, 1958 (6), 11–12). At Kuznetsk, steel is produced in 130-t furnaces with magnesite chrome roofs and is uphill cast into 3·5-t ingots. Aluminium (2 lb/t) is used as a deoxidiser in the ladle. On the basis of formulae given for calculating the optimum amount of Al to be added to the ladle for the deoxidation of non-welding and welding quality tube steel, Al usage for low carbon tube steel was fixed at 1·5 to 1·2 lb/t.—R. S.

The Effect of Aluminium on the Inoculating Power of Ferro-Silicon and its Tendency to Give Pinholing. J. V. Dawson. (*B.C.I.R.A.J.*, 1957, **7**, Aug., 2–9). At least 1% Al is needed for light sections, for heavy sections rather less will serve. The increase in inoculating power of Fe–Si by addition of Al is shown by

the increase of eutectic cells per unit area. A relationship between mechanical strength and no. of cells is demonstrated.

A Contribution to the Problem of the Quality of Rimming Steels. F. Zámečník. (*Hutník (Prague)*, 1958, **8**, Apr., 122–127). The factors governing the appearance of faults in steel ingots with 0·14–0·20% C are examined, and a method is given of calculating the time which should be allowed to elapse between the filling of the mould and the moment when rimming should be stopped. Tables give the results of analyses of ingot faults and their chemical composition. Killing of the steel in the mould is suggested as a method of reducing the scrap and limiting the action of P and S, while reducing ingot growth. For the attainment of optimum results it is necessary to fix exactly the time between the filling of the mould and the completion of rimming. On completion of deoxidation, the surface of the steel must immediately be quenched with water. Further experiments are required to assess the suitability of this method for producing deep-drawing rimming steels.

A Correlation of Data on Deoxidation by Si, Al and Mn with the Nature of Deoxidation Products. Chou Yuan-hsi, Chiang Hsin-yuan and Hsu Kang. (*Acta. Met. Sin.*, 1957, **2**, No. 3, 289–294). [In Chinese]. (13 references).—K. E. J.

The Effect of Furnace Atmosphere due to Combustion Upon Hydrogen Content in Liquid Steel. Lee Hsun and Yang Wei-Shen. (*Acta Met. Sin.*, 1956, **1**, (4), 395–400). [In Chinese].—K. E. J.

An Investigation of Internal Defects in High-Chromium Stainless Steel. Lee Hsun, Ho Chain-chu, Chang Chin-fang, Fung Munchu, Yan Sho and Chao Hwei-tien. (*Acta Met. Sin*, 1957, **2**, (1), 1–12). [In Chinese]. Axial cracking or porosity in ingots and ghost lines in forged billets have the same origins, being associated with the presence of non-metallic inclusions. These defects were reduced by improving the deoxidation procedure during smelting and using thin-walled refractory ingot moulds. Perfectly sound ingots can be obtained by vacuum melting and casting. (29 references).—K. E. J.

"Flocast" Method of Continuously Casting Iron Bar. H. E. Morris. (*Foundry Trade J.*, 1958, **105**, July 3, 5–13). An account of the melting, pouring, pulling and cut-off mechanisms is given. The hardness and other properties of the bar are indicated.

Continuous Casting of High-speed Steel. B. H. C. Waters *et al.* (*J.I.S.I.*, 1958, **190**, Nov., 233–248.) [This issue.]

The Biggest Continuous Casting Plant in the World. V. V. Ful'makht, E. M. Kan and A. G. Chumichev. (*Metallurg*, 1958, (6), 15–17). This plant is under construction at the Kuznetsk plant at Stalinsk. It will cast carbon and low alloy steel from 140-t ladles simultaneously through four strands. Billet and slab sizes range from 4·75 to 10-in. thick and 2 to 4-ft wide. Casting speed is up to 4 ft per min. The machine is in a pit 78 ft deep. Four double-stoppered tundishes are used. Automatic lubrication of the moulds is provided. All processes are automatic. A model of this machine is at the Brussels exhibition.—B. S.

Increasing the Weight of Ingots for Cogging. A. M. Ofengenden. (*Metallurg*, 1958, (6), 13–15). At Kuznetsk six ingots are uphill cast on a bottom plate. It is shown that, by increasing the weight of rimming and killed steel ingots from 2·8 to 3·4 metric t, ingot metal and refractory consumption were markedly reduced and blooming mill productivity was increased.—B. S.

PRODUCTION OF FERRO-ALLOYS

Commercial Grades of Ferronickel and Their Main Uses. J. A. Ternisien. (*Mét. Constr. Mécan.*, 1957, **89**, Nov., 917–925, 940, 953). Various grades of ferronickel are studied, including their elasticity, magnetism and electrical properties. The influence of alloying elements is discussed and suitable compositions for a given purpose are indicated. (31 references).—B. P.

FOUNDRY PRACTICE

Census of Production for Iron Foundries. P. Moore. (*Foundry Trade J.*, 1958, **105**, July 17, 73–75). A discussion of the Iron and Steel Board figures and of the Census of Production 1954, which are used to estimate statistics for the industry as a whole.

Delivery by Pneumatic Tubes Speeds Foundry Tests. (*Iron Age*, 1958, **181**, Apr. 10, 106). Delays in delivering melt samples to the laboratory, and in reporting results were avoided by connecting the shop and laboratory with a pneumatic tube.—D. L. C. P.

[Foundry Practice]. (*Metallografiska Institutets Forskningsverksamhet*, 1956, Jan.-June, July-Dec., 1957, Jan.-June, pp. 26, 27, 24). **The Stability of Moulding Materials of Various Sands.** H. Pettersson. ((**10**), 8–15). This investigation covers grain size, distribution, porosity, tamping, tests with spherical grains, partially controlled grain distribution, tests with blended synthetic sands, and finally conclusions on the results. **Metal Penetration in Box Walls when Casting in Sand Moulds.** H. Pettersson. ((**11**), 13–15; ((**12**), 9–13). **Method of Preparing Collar Cores.** T. Malmberg. ((**11**), 15–19).—F. R. H.

A Simulation of Melting Shop Operations by Means of a Computer. R. Neate and W. J. Dacey. (*Process Control Autom.*, 1958, **5**, July, 264–272).

Preparation and Planning for Modernisation. J. L. Carter. (*Castings*, 1957, **3**, Nov., 47, 49, 50, 51; 1958, **4**, Jan., 33–38; from *Gray Iron News*). Various points needing attention are noted.

Foundry Mechanisation.—III. R. H. Herrmann. (*Foundry*, 1958, **86**, June, 62–65). The permanent mould foundry of the General Electric Co. at Elmira, N.Y. is described. The foundry is equipped with four rotary machines each having 12 moulds for producing grey iron castings. Operating cycle, melting in a neutral-lined cupola and quality control as well as pattern making and storage are described briefly.—T. G.

Automated Foundry System is Flexible. (*Steel*, 1958, **142**, June 16, 98–99). A brief description is given of a system designed for a production jobbing shop, which allows variations in output rate, casting size, shape and weight. Output is 300 molds per h and 36,600 t of castings a year.—A. G.

Aspects of Automation in the Foundry. Examples of its Application. J. Le Covec. (*Fonderie*, 1958, Mar., 105–112). The application of automatic methods to foundry processes is described and illustrated. The problems involved in introducing automation into the foundry are briefly discussed.—B. C. W.

The Problem of Automation in Foundries. P. Schiegries. (*Giesserei*, 1958, **45**, June 19, 363–365). The necessary conditions are outlined for the ability of the foundry to compete successfully with other manufacturing process, the production of baths being cited as an example, with reference to American practice.

Steps to Take in Eliminating Hot Tears. H. Chappie. (*Foundry*, 1958, **86**, July, 82–84). Hot tears in a press platen weighing 27,000 lb. with ribs up to 22 in. high were eliminated by design change and appropriate foundry practice, e.g. using a sand which absorbs expansion (wood flour); multiple gating of correct thickness; use of external chills; favourable design (staggered ribs); and use of an iron with 0·40% C. After employing this technique, welding time for repair of the castings was reduced from 50 to 1 hr.

The Cleaning and Finishing of Heavy Steel Castings. H. C. J. Corke. (*E.S.C. News*, 1958, **11**, Spring, 4–9). An illustrated account of the various operations.

How to Set Production Standards for Pouring, Shifting, Shakeout. J. E. Keith. (*Foundry*, 1958, **86**, July, 85–89). The author describes a simple, accurate, and inexpensive method of setting production standards for pouring, shifting, and shakeout operations. A number of examples are quoted and some forms for calculating efficiency illustrated.—T. G.

Close Control Keeps Foundry on Schedule. G. E. Loftin. (*Iron Age*, 1958, **181**, May 8, 100–103). The process planning and control system at a jobbing foundry are described. The method gives priorities to the work and loads the plant, and has considerably reduced the number of late deliveries.—D. L. C. P.

Changing Physical Demands of Foundry Workers in the Production of Medium Weight Castings. H. Scholz. (*Ergonomics*, 1957, **1**, Nov., 30–38). Studies on conditions in 7 foundries are reported, both old-fashioned and mechanized. High physical demands were found to exist even in the latter. The various trades are considered and performance levels and rest periods are shown.

VACUUM METALLURGY

Instrumentation in Vacuum Induction Melting. (*Ind. Heat.*, 1958, **25**, Apr., 732–734). The control system of an industrial furnace of one ton capacity is described.

Laboratory Arc Furnace for Argon and Vacuum Melting. (*Brit. Steel*, 1958, **24**, Apr., 110–111). A new laboratory arc furnace introduced by Edwards High Vacuum Ltd. is capable of quick adaptation for either consumable-electrode or non-consumable electrode melting, and either under an A atmosphere or in a vacuum. Details are given of the design of the unit and its pumping system.

New High Temperature Furnace. (*Iron Steel*, 1958, **31**, Mar., 112). Graphite resistor furnaces capable of providing temperatures up to 3000° C have been developed by G.E.C. Ltd. Brief details are given of a unit installed at the B.S.A. Laboratory in Birmingham for sintering materials for use at high temperatures. It is designed for temperatures up to 2000° C and to maintain a good vacuum to prevent oxidation of the charge.—G. F.

Study on the Properties of Vacuum-Melted Steel. I. On Quench Age-Hardening. R. Yamada, K. Yokoyama, T. Nakamura and F. Urushibara. (*Nippon Kinzoku*, 1957, **21**, Mar., 141–145). [In Japanese]. A range of carbon steels melted *in vacuo* and in the atmosphere, containing carbon in the range 0·02–0·87%, and of commercial steels, was examined. The as-annealed hardness and the as-quenched hardness of the vacuum-melted steels were lower than those of the steels melted in the atmosphere and the commercial steels; the hardness increase and increasing rate of hardness due to ageing were also lower. The change of hardness due to ageing in vacuum-melted steels was most marked when specimens containing 0·05–0·1% C were quenched from approx. 700° C.—K. E. J.

Vacuum Pouring of Ingots for Heavy Forgings. J. H. Stoll. (*Blast Furn. Steel Plant*, 1958, **46**, June, 595–605). The author describes two 250-t vacuum degassing units for ingot casting. On top of the vacuum unit is mounted a pony ladle which is filled from the furnace ladle. The vacuum equipment is described as well as the monitoring system. The latter consists of two TV cameras and a high-speed cine camera. Preparation of tank and mould, operation of the vacuum unit, pouring stream, hydrogen removal, the properties of the steel, including defects, and finally cost factors are dealt with.—T. G.

REHEATING FURNACES AND SOAKING PITS

Convection and Radiation in High Speed Heating. N. H. Davies and R. J. Reed. (*Steel Proc.*, 1958, **44**, Mar., 155–160, 162). Soaking pit and heat treatment furnace conditions are discussed. (21 references).

Resistance Heating of Billets. W. Trinks. (*Ind. Heat.*, 1958, **25**, April, 712–714). A summary of the paper by W. Stich, *Stahl u Eisen* 1954, April 4.

New Continuous Bloom Reheating Furnace. (*Instr. Practice*, 1958, **12**, July, 784–785). A furnace with a hearth 75 × 16 ft installed at Cargo Fleet for the Ormesby Rolling Mill is described with special reference to control.

Numerical Evaluation of Temperature Distribution in Forging Ingots. N. V. Hunt. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 25–31). A

procedure for predicting the conditions in forging ingots during soaking is described, with an example.

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Recent Heat Treatment Furnace Installations. (*Metallurgia*, 1958, **57**, June, 283–302). This comprehensive review shows a trend towards furnaces of greater capacity and simplified operation. Improvements are in detail only. Greater attention is being given to the problem of controlled atmospheres.—A. G.

Choice of a Furnace for Thermal Treatments. V. Lalonde. (*Chal. et Ind.*, 1958, **39**, May, 130–134). Various factors influencing the choice of a heat-treatment furnace are enumerated and discussed.

Definitions of Terms Used in the Heat Treatment and Testing of Metals. (*Australian Institute of Metals publication*, 1957, pp. 88).

Automatic Control of Gas-Fired Furnaces. L. Walter (*Gas World*, 1958, **38**, July 19, 8–9). Factors in control are briefly reviewed.

Two Steps to Heat Treat Savings. (*Steel*, 1958, **142**, June 16, 100–101). Changes in the methods of fabricating high strength automotive steering knuckles and of quenching transmission gears are described which have led to cost savings.—A. G.

An Application of the Theory of Economic Loads to the Calculation of Heat-Treatment Furnace Capacities. G. Auke Bea. (*Dyna*, 1958, **33**, May, 354–355). [In Spanish]. An equation is derived that relates the economic loading of a furnace to the physical and economical factors of its operation. A numerical example is given.—F. S.

A Study of Temperature Uniformity in Heat Treating. H. N. Ipsen. (*Met. Prog.*, 1958, **73**, May, 89–92). Data are given which show that fans and baffles promote uniformity of temperature when packed charges are heated in a muffle furnace. The importance of accurate temperature distribution in gas carburizing and in the treatment of alloy steels is stressed.

Contribution to the Industrial Application of TTT Diagrams to High Cooling Rates in the Heat-Treatment of Steels. H. Bühler and F. J. Schmidt. (*Stahl Eisen*, 1958, **78**, May 15, 663–638). *TTT* diagrams show the structure of a steel after slow cooling to room temperature, and are usually not applicable to materials that have been rapidly cooled. The authors quenched steel bars of up to 400 mm diameter in oil and water and recorded the cooling curves. They found that the results obtained were in good agreement with calculations by the Russell–Bachmann method for bars up to 500 mm in diameter.—T. G.

Decarburization of White Irons under Non-Graphitizing Conditions. C. T. Moore. (*B.C.I.R.A.J.*, 1958, **7**, June, 258–261). Excess S white irons decarburize more slowly than balanced S irons before graphitization begins. Increase of the austenite–carbide interface increases the rate of decarburization in the initial stages. The results are consistent with the suggestion that carbide solution in austenite is a limiting factor influencing C removal. Free FeS also seems to reduce the rate of solution of carbide in austenite.

Studies on Heat-Treatments of High Alloy Tool Steels. III. On Stepped-Quenching of High-Carbon High-Chromium Steel. T. Nemoto. (*Nippon Kinzoku*, 1957, **21**, Feb., 89–92). [In Japanese]. Results were obtained from hardness, impact, bending and magnetic measurements. The amount of retained austenite in stepped-quenched is more than in normally quenched steel, and mechanical properties are improved. Hardness and retained austenite decrease with increase of holding time in the hot bath (500° C). Retained austenite increases markedly with decrease of the cooling rate from 500° C. (22 references).—K. E. J.

On the Heat-Treatment of a Alnico-5 Magnet. T. Mishima, R. R. Hasiguti and Y. Kimura. (*Nippon Kinzoku*, 1957, **21**, Mar., 187–190). [In Japanese]. The coercive force (or residual induction) of Alnico-5 changes logarithmically with cooling rate when cooling from 1000° to 700° C, with or without a magnetic field. Changes of coercive force as a function of ageing time after cooling in a magnetic field can be expressed by the Austin-Rickett formula. The results are explained qualitatively in terms of structures and activation energy. (11 references).

The Influence of Design and Quenching Technique on Heat Treatment Stresses. A. G. Gardner. (*Wild Barfield J.*, 1958, **6**, June, 2–5). Effects of shape and modified quenching methods are briefly indicated.

Hot-Bath Quenching at Low Temperatures in Eutectoid Carbon Steel. II. On the Relation between Toughness and Structural Constituents. I. Tatsukawa. (*Nippon Kinzoku*, 1957, **21**, Apr., 203–206). [In Japanese]. Results are given for 0·8% C steel wires hot-bath quenched both above and below the Ms temp. The shapes of the curves for absorbed energy for fracture in the bending test against holding time in isothermal treatment are similar whether the soaking temp. is above or below the Ms point; every curve has a max. corresponding with a structure giving improved toughness. Retained austenite decreases the yield stress; and superior toughness with high hardness can be obtained by certain hot-bath quenching in the martensitizing range, as compared with normal oil-quenching and tempering.—K. E. J.

"Structural Shrinking"—A New Shrinking Process. A. Peiter. (*Werkstalltechnik Maschinenbau*, 1958, **48**, May, 265–270). A process is described, by which changes in shape of hardened steel rings on heat treatment are utilized for shrinking on the rings. By studying the resultant stresses the influence of the temperature of heat treatment and the geometry of the rings are examined, and the results discussed.

On Repeated Subzero Treatment of Steel. M. Kondo and T. Hachisuka. (*Nippon Kinzoku*, 1957, **21**, Apr., 227–230). [In Japanese]. Changes in hardness and length of high-carbon low-chromium steel are reported to show the effects of quenching temp., martempering and tempering on repeated subzero treatment at varying temp.—K. E. J.

Furnace Design for Hardening and Tempering. (*Ind. Gas*, 1958, **21**, July, 16–18). Furnaces designed by the Gas Light and Coke Co. for continuous treatment of small parts are described.

Studies on the Temperability of Carbon-Steel Wires (2nd Report). On the Effect of Carbon Content. M. Sugiyama, (*Nippon Kinzoku* 1957, **21**, Aug., 498–591). Wires of 0·62, 0·80 and 1·08% C were quenched and tempered at various temperatures, and their mechanical properties assessed by behaviour in a bend test. The critical tempering temperature, above which the specimens did not break, increased with carbon content. The 1·08% C steel had the highest strength when it was austenitized non-uniformly. Generally, the critical temperature was clearly defined, in terms both of percentage of specimens fractured in the bend test and of angle of bend at which fracture occurred.—J. G. W.

Study on "Strain Tempering (KRK Process)." IX. On the Changes of Anisotropy in a Silicon–Iron Alloy by Strain Tempering. M. Kyotani. (*Nippon Kinzoku*, 1957, **21**, Mar., 163–166). [In Japanese]. Results are reported for 2·6% Si–Fe alloy strip. The magnetic characteristics are effectively improved by the KRK process, and it is most effective when strain-tempering is applied after the skin pass of a few per cent. reduction. The treatment accelerates the rotation of crystalline orientation to the direction of magnetization.—K. E. J.

FORGING, STAMPING, DRAWING AND PRESSING

Stainless Forging Takes Precise Control. (*Steel*, 1958, **142**, June 9, 92–94). Techniques are described for the successful forging of stainless steel parts. Proper cutting and handling minimise carburization and splitting.

Big Drop Hammer Raises Quality, Lowers Cost. G. H. DeGroat. (*Metalw. Prod.*, 1958, **102**, Aug., 1, 1350–1351). A 50,000 lb steam hammer at Park Drop Forge Co. for closed-die forging techniques is used for heavy forgings which are discussed.

3 New Forging and Extrusion Methods. J. F. Murphy. (*S.A.E.J.*, 1958, **66**, June, 40–42). A precision gear forging process, a rotary forging process, and the cross-extrusion process (for springs) are described. The uses and advantages of each are briefly noted.

Hot-Pressing with H.F. Induction Heating. R. Schiesser. (*Metalw. Prod.*, 1958, **102**, June 27, 1123–1126). A Brown-Boveri plant and its operation are described. Small parts are produced.

Tool Wear in the Cutting and Blanking of Sheet Steel. O. Kienzle and W. Kienzle. (*Stahl Eisen*, 1958, **78**, June 12, 820–829). The cutting and blanking mechanisms are explained and studies on blanking using various tool steels and a variety of sheets are reported.

Steel-Tube Stampings: Short Cut to Complex Shapes. J. F. Hullman. (*Iron Age*, 1958, **181**, Mar. 6, 122–125). Examples of instances where savings have been made by making parts from welded steel tube instead of solid bars or castings, are given.—D. L. G. F.

Dinking Dies Cut Blanking Costs on Thin Sheet. F. Strasser. (*Iron Age*, 1958, **181**, June 12, 90–91). The design and methods of use of dinking dies are described. These are inexpensive and suitable for blanking or piercing thin sheets (up to 0·08 in. thick) of soft metals.—A. G.

Induction Heating Facilitates Axle Production. (*Ind. Heat.*, 1958, **25**, Apr., 720–724). Drop axles are produced by the machining of bar stock followed by bending after induction heating, instead of from drop forgings.

The Possibilities of Developing a Non-Ageing Deep-Drawing Steel. D. Tlustá and J. Skála. (*Hutnik* (*Prague*), 1958, **8**, Apr., 118–122). The subject is written up from West European and U.S. sources. In Czechoslovakia the production of deep-drawing sheet represents a considerable proportion of total sheet production, but in fact it's quality leaves much to be desired. The problem of its improvement is under attention, and it must be admitted that the method of stabilization of rimming steels, taken up experimentally in Czechoslovakia about a year ago, is now confirmed as correct in the light of world trends and of the results of these experiments to date.—M. T.

Increasing the Strength and Wear Resistance of Machine Parts by Roller Burnishing. A. I. Barats. (**Metallov. Obrabotka Met.*, 1957, (9), 47–51). Equipment used and the performance of the hardened parts is described. A unit for surface hardening parts up to 205 mm diameter has been constructed. It is set up on a lathe bed and travels along the workpiece axis.

"Explorer" Nose Cold Formed From 430 Stainless. (*Iron Age*, 1958, **181**, Mar. 6, 113–115). The 'Floturn' cold roll-forming technique was used to shape the nose-cone of the Explorer earth satellite. The cone is 12 in. long, 6 in. dia. at the base and approx. 0·015 in. skin thickness, increasing to 0·094 in. at the tip. In the 'Floturn' process a roller-type tool spreads a metal blank to the shape of a mandrel, as the workpiece and mandrel both revolve. Considerable benefit accrues from the extreme degree of cold work.—D. C. L. F.

Explosive Metallurgy. H. P. Tardif and W. H. Erickson. (*Can. Min. Met. Bull.*, 1958, **51**, June, 352–359). A review of the deformation of metals under shock, the markings produced on the surfaces and the shaping of metals by means of explosive charges is given. (40 refs.)

Wire Drawing Plant. (*Mech. World*, 1958, **138**, July 305–307). An account of the Kingfield works installation. Equipment is listed and heat treatment outlined.

Friction and Lubrication in Wiredrawing. A. E. Jones. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 13–20). Dies are also discussed.

Effect of Working Conditions in Patenting of Wire Rope by Resistance Heating on the Mechanical Properties. W. Lueg and K. Schemmer. (*Stahl Eisen*, 1958, **78**, July 10, 960–965). Patenting by resistance heating

as compared with patenting in a conventional gas-fired oven gave better tensile and bending properties of the steel wire, the properties of the resistance-heated wire were more uniform. The results are presented in many graphs.—T. G.

Transformation Phenomena Occurring in the Patenting of Wire Rope by Resistance Heating in Relation to Strength Properties. A. Rose, L. Rademacher and K. Schemmer. (*Stahl Eisen*, 1958, **78**, July 10, 966–974). The metallurgical processes occurring in the patenting, and the formation and disintegration of austenite were studied by thermal analysis and dilatometric tests. *TTT* diagrams are presented. The transformation mechanism during patenting is explained. It is shown that the absolute tensile properties of the wire are independent of the type patenting.—T. G.

ROLLING-MILL PRACTICE

A New Theory of Rolling. A. Geleji. (*Acta Techn.*, 1957, **19**, (1–2), 199–243). [In German]. A critical examination of existing theories is made and defects in the assumptions behind them are indicated. Enlargement and forward slip are then taken into account and the forces involved, including those previously neglected, are now considered. Compressive stress at any point between the rolls which with the yield strength of the material make up the system, is claimed to be made up of two components one from the frictional forces and the other from the extruding effect of pressure in an axial direction giving forward slip. The expressions produced hold for both hot and cold rolling.

The Two-Flank Rolling Diagram and Errors in the Operational Condition. F. Krüger. (*Werkstattstechnik Maschinenbau*, 1958, **48**, May, 259–261). In certain cases the two-flank rolling diagram can indicate excessive errors. The cause of these is discussed theoretically and with reference to examples.

On the Hot Workability of Low Carbon Special Shaped Steel. T. Akutagawa, Y. Mishima, E. Asano and A. Suseki. (*Ann. Rep. Eng. Res. Inst. Tokyo Univ.*, 1957, **16**, Sept., 75–80). [In Japanese]. Surface crazing defects observed at 850–1250° C during rolling depend on the amounts of Cu, Sn and S present and on factors concerning the refining and deoxidizing periods. (14 references).

Two-Dimensional Distribution of Friction Components for Plate Rolling. Wang Meiying. (*Acta Met. Sin.*, 1956, **I**, (3), 295–300). [In Chinese]. The surface distribution of friction components is calculated using a method based on the "rule of gradient" and an experimental pressure surface. The surface obtained gives a boundary condition for the case of three-dimensional plate rolling, and the method is also applicable to forging etc. (10 references).—K. E. J.

Is Your Lube Oil Giving You Top Mileage? (*Iron Age*, 1958, **181**, Feb. 27, 108–109). The operation of the airborn-fog centralized lubrication system on a 700 ft rod-cooling unit is described. The "Micro-Fog" system serves 576 bearings, 188 gears and 6 drive mechanisms using 1¼ quarts/h oil.—D. L. C. P.

Roll Pass Design in Section Mills. I. F. Revenko and I. M. Konovalov. (*Metallurg*, 1958, (6), 23–25). The authors discuss improvements in pass design for spring steel bars in the 16 in. mill, for hexagons and squares in the 14 in. mill and for $\frac{5}{16} \times \frac{1}{2}$ in. in the 10 in. mill at Kuznetsk.—R. S.

Light Section Rolling Mills. S. Makrlik. (*Czech. Heavy Ind.*, 1958, (6), 25–30). A general account of Czech installations and practice.

Roller Guides for a 10-in. Rod Mill. B. M. Maksimov, N. P. Ghetvin, A. A. Ivanov and G. V. Babkov. (*Metallurg*, 1958, (5), 28–30). Guides of Cr-steel are described. They have a life of 40–45 shifts rolling 8–12 mm rounds in the second train and a life of 18–20 shifts rolling 5·25–7 mm rounds in the third.—R. S.

Rolling Mills. V. Matoušek. (*Czech. Heavy Ind.*, 1958, (6), 20-25). A general account of Czech installations and practice.

The Planetary Mill and its Uses. T. Sendzimir. (*Iron Steel Eng.*, 1958, **35**, Jan., 95–99). The operation of the mill is described in detail, and various types of complete installation are discussed. The strip temperature is normally higher at exit than at entrance to the mill; hence obvious applications are to rolling hard metals, or metals where scaling is to be avoided. Precise temperature control can be achieved with slabs butting against one another, and protective atmospheres can be used. A vertical mill to work with continuously cast slabs is being designed.

Effect of Working Conditions of a Three-Stand Cold-Rolling Tandem Mill on the Mechanical Properties of Strip Steel. J. G. Brockhaus, F. Fischer, H. Graetz and K. Ohlwein. (*Stahl Eisen*, 1958, **78**, July 10, 955–960). The total output of the mill depends for equal total reduction on the selection of the individual reductions in the three stands. The total output is highest for high reductions in the second and third stands and for low reductions in the first stand. High reduction in the last stand yields high tensile strength in the strip. Tests on lubrication show that tallow emulsions are best. Using tallow emulsions the relation of yield strength, tensile strength and hardness remains unaltered after annealing and a finishing pass.

The New Hot-Strip Rolling Mill of Borsig A.G., Berlin-Tegel. W. Stenger and O. Dahlke. (*Stahl Eisen*, 1958, **78**, June 12, 812–820). The steelworks at Borsig, Tegel, was supplied with a hot-strip mill for supplying their tube plant. The mill consists of a two-high reversing stand and of five, four-high, finishing tandem stands. The ingots are rolled into the finished strip in one operation without intermediate reheating. The surface quality of the strip is sufficient for tube fabrication.—T. G.

The Production of Steel Strip and Sheets with Particular Reference to Research and Development. H. Edwards. (*Trans. Manchester Assoc. Eng.*, 1955–1956, 355–377, discussion 378–386). An outline of all stages of the process is given, with a section on quality control.

Effects of Finishing and Coiling Temperatures on the Properties of Hot-Rolled Steel Strip. J. K. Macdonald. (*B.H.P. Tech. Bull.*, 1958, **2**, Apr., 38–42). Grain structures of rimmed and semi-killed strips are shown.

Lay Out and Regulation of Cold Rolling Mill. H. Brunsberg. (*La Technique Moderne*, 1958, March, 92–98). The problems involved in selecting electrical equipment for cold rolling mills are treated. A study is made of systems with multiple motors and equipment for total control is described.

The Cold Rolling of Sorbitic and Pearlitic Steels of High-Carbon Contents. W. Roth, H. Lassek and F. Fischer. (*Stahl Eisen*, 1958, **78**, July, 10 948–955). Cold rolling of sorbitic and pearlitic steels of high-carbon content (0·62–1·20%) at various reductions per pass showed the excellent deformability and high U.T.S. attained by sorbitic steels. Cold rolling reduces the transformation temperatures and holding times in intermediate heating.—T. G.

Reversing Rolling Mill Drives. (*Metallurgia*, 1958, **57**, June, 309–311). Requirements for large industrial reversing drives and the advantages of using mercury arc convertors for this purpose are discussed.—A. G.

Electrical Equipment for a Modern Rod Mill. E. L. Anderson. (*Iron Steel Eng.*, 1958, **35**, Jan., 112–123). Full details are given of control and drive equipment on the rod mill at the Johnstown plant of Bethlehem Steel Corp.—a 23-stand, 3-strand, high-speed mill, having nine roughing stands, six intermediate stands and eight finishing stands. The eleven motors give 7450 h.p. total.—K. E. J.

New Methods for the Selection of Personnel for a Modern Cold-Rolling Mill. W. Fuchs. (*Stahl Eisen*, 1958, **78**, July 10, 974–977). The author describes the psycho-physical requirements of personnel for key jobs in a modern cold-rolling mill and the various tests that were carried out in the selection of applicants. After two years' operation of the mill it was ascertained that the methods used for the selection were correct.—T. G.

Measuring Width in a Hot Strip Mill. (*Metallurgia*, 1958, **57**, June, 307–308). A brief account is given of a gauge for measuring width in a hot strip mill. The gauge, based on a system put forward by BISRA, utilizes infra-red radiation to actuate a servo-mechanism. Accuracy is about ½in.—A. G.

A Digital Flying Extensometer for Temper Rolling Mills. N. S. Wells. (*Trans. A.I.E.E.*, 1957, **76**, Part 2, 376–378). Decade counters are attached to entry and exit tension rolls and extension is determined directly to ± 0·01% or better.

Instruments for Recording Rolling Processes. E. I. Sametskii. (*Metallurg*, 1958, (6), 30–31). Instruments for recording the number of passes, absolute reduction at each pass, time taken for each pass, data on manipulators, etc. in a 3-high reversing mill are described. Instrumentation of a blooming mill which records time from arrival of an ingot to completion of the last pass, number of ingots cogged, and rolling temperature of each ingot is also reviewed.—R. S.

Trucks Speed Coil Drying. (*Steel*, 1958, **142**, May 19, 146). In the installation mentioned, trucks carry the wet coils through direct gas-fired ovens on a time schedule.—A. G.

MACHINERY AND SERVICES FOR IRON AND STEEL PLANT

Steelworks Power Station at Ravenscraig. W. M. Kerr. (*G.E.C.J.*, 1958, **25**, Apr., 122–126).

Two Gas-Turbine Sets in a Belgian Iron Works. A. Jaumotte. (*Stahl Eisen*, 1958, **78**, May 15, 653–663). Design and layout of two gas-turbine sets each of 7500 kW output are described. When the turbines were ordered in 1952, it was thought that a dust content of 20 mg/m³ of the flue gas could be permitted. It was found that frequent cleaning of the turbine and the coolers is necessary. A reduction of the dust content to 1–4 mg/m³ is recommended.—T. G.

Instrumentation for Turbo-blower: Kent Equipment at Appleby-Frodingham. (*Iron Steel*, 1958, **31**, May, 193–194). Details are given of the Kent instrumentation and automatic control equipment installed on the turbo-blowers at the Appleby-Frodingham blast-furnace plant. The method of operation and the accuracy limits attained are discussed.

New Scrap Baler: Triple Action Fielding Machine Shows its Paces. (*Iron Steel*, 1958, **31**, Mar., 105–106). A new triple-compression hydraulic press scrap baler made by Fielding & Platt Ltd. is a completely self-contained unit capable of handling up to 20t/h of heavy grade scrap metal. A description of the unit is given.—G. F.

Radio Speeds Scrap Handling. (*Iron Age*, 1958, **181**, Apr. 17, 106–107). The plant of a large scrap-yard and the control methods used, are described.—D. L. C. P.

Conveyors for Swarf Handling. (*Mech. Handling*, 1958, **45**, Aug., 524–527). The Standard Motor Co. Banner Lane works installations are described.

"Non-Fixed" Conveyors and Elevators. J. M. Beskine. (*Mech. Handling*, 1958, **45**, Aug., 528–537). Machines handling furnace coke and other materials are described.

Rail Transport Aids Britain's Record Iron and Steel Production. M. D. J. Brisby. (*Transport Age*, 1957, **1**, Oct., 19–23). A general account of internal and external rail transport for the iron and steel industry.

WELDING AND FLAME-CUTTING

Welding—Is It a Menace? An Engineer's View. F. Koenigsberger. (*Trans. Manchester Assoc. Eng.*, 1955–1956, 83–109, discussion 109–120). A review of welding defects, of its use in the construction of equipment and choice of method and of proper procedure in fabrication and repair.

Welding in the Steel Foundry. O. Bergmann and H. Reichert. (*Giessereitechn.*, 1958, **4**, May, 110–111). The use of welding

in the steel foundry is discussed, and two examples are given of the repair of castings by welding.

The Repair of Steel Castings by Welding. L. Müncner. (*Slévárenství*, 1958, **6**, Apr., 106–109). The author outlines what a welder requires to know about the chemical composition, above all the carbon content, and the physical properties of steels, when he is called upon to repair worn castings or make good defects in new castings by electric arc welding, in order to choose the appropriate electrode and decide what pre- and post-weld heat treatments are required. The economic advantages of the use of electric arc welding are indicated.—M. T.

Welding: Corrosion-Resisting Steels of the 18/8 Cr/Ni Type and Heat-Resisting Steels. F. A. Ball. (*Sheet Metal Ind.*, 1958, **35**, July, 509–522). The use of the atomic hydrogen torch, arc welding and the electrodes used, argon-arc, submerged arc and resistance welding are reviewed. A note on sigma-phase is added. (28 references).

Welding of Nickel and High-Nickel Alloys. J. Hinde. (*Brit. Weld. J.*, 1958, **5**, July, 311–318). Clad steels are included.

The Welding of Thick Steel Plates. (*Atomics Nucl. En.*, 1958, **9**, Apr., 124). A note on the use of the Fortrex 35A electrode in the welding of 3–4 in. plate for nuclear reactors.

Welded Air-Cleaner Assembly Makes for Big Savings. W. E. Meagher. (*Iron Age*, 1958, **181**, Feb. 27, 110–111). Advantages of welded steel compared with soldered cast iron for air cleaner units are described. Weight has been reduced from 26¼ lb to 11½ lb, and leaks almost eliminated. The cost was reduced 36%.—D. L. C. F.

Cutting and Welding of Petroleum Refinery Equipment. C. B. Robinson. (*International Acetylene Assoc. Proceedings*, 1956, 38–42).

The Uses of the Oxy-Acetylene Flame in the Maintenance-of-Way Department. T. S. Bean. (*International Acetylene Assoc. Proceedings*, 1956, 26–31). Railway uses are mentioned including metallizing and pretreatment and finishing of crankshafts and spray welding of cracks in cylinder heads and water jackets.

Welding of Clad Steels. A. Franzén. (*ESAB Rev.*, 1957, (2), 3–9). [In French]. Stainless-steel clad specimens were welded. Choice of welding rod is discussed and the various methods of making the welds and their properties are described.

Metallurgical Considerations in the Welding of Cladded Steel Sheet. W. Rädeker. (*Schweissen Schneiden*, 1958, **10**, July, 255–259). The welding of ferritic Cr and austenitic 18/8 Cr–Ni steels clad with Ag, Cu or Ni or with Monel is described. Recommendations for heat-treatment after welding are included.

How and Where to Use Shielding Gases for Welding. O. T. Barnett. (*Iron Age*, 1958, **181**, Apr. 10, 91–98).

Operation Face Lift. (*INCO Mag.*, 1958, **27**, Apr., 6–7). Rebuilding railway carriages with extensive use of stainless steel at a cost 15% that of a new carriage is described. Welding is widely used.

Electro-Slag Welding in the Production of Boiler Shells. V. I. Rabinovich. (*Weld. Met. Fab.*, 1958, **26**, Apr., 134–137, 156). In the electro-slag welding process, recently developed in Russia, a filler wire is melted directly into a "mould," two walls of which consist of the edges to be joined and the other two walls of specially shaped water-cooled slides. Some operational data of the process are given.—G. F.

Ultrasonic Welding Makes Rapid Advances. (*Steel*, 1958, **142**, Mar. 31, 80–81). Examples of developments in ultrasonic welding techniques are described. Strong bonds in Al and stainless metals are possible. Seam welding and the joining of odd shapes are fields of advance.—D. L. C. F.

Ultrasonic Welding—A New Technique Grows. J. B. Jones. (*Met. Prog.*, 1958, **73**, April, 68–72). The method is used to join similar or dissimilar metals including stainless steels in thin gauges. The joint is as strong as those produced by conventional methods, but no fusion occurs.

Electrode-Tip Life Studies in Series Spot Welding. E. F. Nippes, W. F. Savage, S. M. Robelotto and K. E. Dorschu. (*Weld. J.*, 1958, **37**, June, 241s–248s). The welding of rimming steel sheet is examined for the effects of electrode shape, material and welding current on weld strength and consistency and on electrode life.

The Great Influence of Welding on Technology. K. Klöppel. (*Schweissen Schneiden*, 1957, **9**, Dec., 503–512). A wide review of welding techniques applied to structures and in reactor construction, with a brief reference to welding classification.—U. E.

Inert-Gas Tungsten-Arc Welding SAE 4130 Steel Sheet—3. The Effects of Welding Speed on Cracking. C. A. Terry and W. T. Tyler. (*Weld. Met. Fab.*, 1958, **26**, Jan., 24–31). Welding and hot-cracking tests are reported and hardness and structures are examined. Cracking occurs in a limited range of welding speeds and the form of the relationship of cracking and welding speed whether designed to produce cracks or not, is so similar that the cracking observed must be hot-cracking. Ferrite-pearlite banding has a marked effect on cracking severity and the pearlite may remain fluid longer than the pearlite, producing a notch effect.

Developments in Stainless-Steel Welding in the Nuclear Program. E. B. La Velle, L. H. Rasmussen and E. M. Kuchera. (*Industrial Atom. U.S. Atomic Energy Commission Tech. Inf. Service*, 1956, Aug., TID–8013, pp. 19). Purging is done with A, the joint is tacked securely at a spacing slightly less than the filler rod diameter, the rod is held down tightly into the joint root and at a tangent to the pipe at that point, the arc is concentrated on the end of the rod at the point where it is held against the root joint, the arc is kept as short as possible (⅛ in. appears to be best), and is traversed along the joint to form the root closure weld. Typical joints are shown.

Welding of Plastic-Coated Steel. D. Ewan. (*Weld. Met. Fab.*, 1958, **26**, Jan., 22–23). Spot and seam welding methods are indicated.

The Pressure Welding of Iron-Carbon Alloys in the Range 0·50–1·9% Carbon. J. F. Robinson and R. F. Tylecote. (*Brit. Weld. J.*, 1958, **5**, Feb., 77–81). Pressure weldability in a H_2 atmosphere was examined at 805–1105° C. The steels contained 0·48%, 0·83% and 1·88% C and the presence of other elements was avoided as far as possible. Tensile tests were made on the welds, which were also examined metallographically. All welds were satisfactory.

Cold Pressure Welding and Brazing. F. C. Fritz. (*Werks. Korr.*, 1958, **9**, Jan., 4–7). Joining of Armco iron and steel are included. The process and methods used are described.

Stress Relief of Weld Layers Assures Quality Repairs. (*Iron Age*, 1958, **181**, Jan. 30, 98–99). The repair of a fractured 5-t crankshaft is described. Alternate welding and stress relieving, taking 150 man-hours, gave an effective repair at one fifth of the replacement cost.—D.L.C.F.

Induction Heat Treatment of Welded Joints of Carbon Steel. V. V. Aleksandrov and T. A. Vaskchurova. (*Autogennoe Delo*, 1949, **20**, (8), 20–23). A study of heat treatment leads to the conclusions that prolonged holding at a high temperature is not necessary, as quick heating to 650° C and air cooling is sufficient. An inductor design is given for circular or longitudinal welds in boilers and petroleum crackers and considerable improvements in output and power consumption can be obtained.

Welding Properties of Hard and Alloy Steels. E. Henrion. (*Prat. Soud.*, 1956, **10**, Oct., 156–161; Nov., 183–188; Dec., 206–210; 1957, **11**, 9–14). A review in general terms of the properties and structures of mild and hard carbon steels and effects of pre-heating and other practices on hardness and other properties of welds. Methods of flame and electric welding are reviewed showing the use of various types of rod coatings, and a summary is added.

Influence of Gases in Arc Welding of Steel. J. D. Fast. (*Schweissen Schneiden*, 1957, **9**, Dec., 512–517). The physico-chemical reactions which occur during welding with coated electrodes are discussed. The interaction between metal and gas is investigated. Causes of porosity in weld metal are discussed together with the harmful effect of gases in solid metal.—U. E.

Effect of Hydrogen on the Properties of Welds. P. C. van der Willigen. (*Schweissen Schneiden*, 1957, **9**, Dec., 517–521). The effect of hydrogen in weld metal is investigated. A review is given of work carried out on micro- and macro-cracking in welds due to the presence of H_2 during arc welding and the welding of mild steel. The formation of fish eyes in rolled steel samples is discussed.

A New Brazing Technique. C. E. Rorick. (*Weld Met. Fab.*, 1958, **26**, Jan., 19–21). Production of honeycomb panels brazed with 85% Ag–15% Mn alloy operating up to 1000° F is described.

Brazing Strip Eliminates Rejects. (*Steel*, 1958, **142**, Feb. 10, 83–84). Prediffused silver brazing alloy on ferrous or non-ferrous strip gives satisfactory joints.

The Metallurgical Reactions between Brazing Alloys and Base Materials and Resulting Conclusions Appertaining to Constructions Designed for Brazing. J. Schatz. (*Schweissen Schneiden*, 1957, **9**, Dec., 522–530). The thermodynamics of brazing reactions are investigated. It is shown that wetting of a solid metal by means of a liquid solder can be determined by the free energy ΔG of alloy formation occurring at the boundary. Effects of gases for shielding and of fluxes are discussed.—U. E.

The Use of Gas for Brazing. (*Techn. Mod.*, 1957, **49**, Sept., 484–488). Various types of brazing processes using gas heating are described. The aspects covered include brazing torches, furnace brazing, and high temperature radiant burners.—B. C. W.

Bibliography on Adhesive Bonding of Metals. Engineering Societies Library. (*ESL Bibliography*, No. 12, pp. 23). The period covered is 1947–1957 with over 150 references to books and periodicals.

Three New Adhesives for Steel, Other Materials. (*Mat. Design Eng.*, 1957, **46**, Dec., 156–157). A cyanoacrylate and two epoxy compounds are noted.

Liquid Adhesive Bonds Steel. (*Iron Age*, 1957, **180**, Nov. 7, 136). The properties of a new liquid adhesive "Eastman 910" are described. A rapid set-time and high strength are obtained on a variety of materials. The adhesive action results from the rapid polymerization of a cyanoacrylate monomer.

The Influence of Welding Current on the Economies and Quality of Arc-Welded Butt Welds. C. Stieler. (*Schweissen Schneiden*, 1958, **10**, April, 119–125). It was found that with increasing welding current and increase in electrode diameter the rate of fusion increased considerably while welding time decreases. A series of tests carried out on St 37 using a current recommended by the electrode manufacturer and 20A either above or below the recommended value showed that the tensile strength was very little affected by it except with one electrode of 4 mm dia. The notch impact values decreased with increase in current.—U. E.

The Load-Carrying Capacity of Silver-Brazed Joints for Thin Walled Steel Tubes. H. Koch and K. L. Poth. (*Schweissen Schneiden*, 1958, **10**, April, 114–119). Silver-brazed (40% Ag) tube sections of St. 35.29 gave a strength value of 40 kg/mm² as compared with 56 kg/mm² for cold-drawn tubes, brazed tube sections in cross form gave 25 to 30 kg/mm² while the shear strength of the brazed seams of cross samples was 16–20 kg/mm².—U. E.

The Ultrasonic Testing of Welds. M. J. Cotter. (*Atomics Nucl. En.*, 1958, **9**, Apr., 115–117). A general account of ultrasonic testing is given with a list of commercial equipment for weld examination.

The Radiographic Examination of Welds in Mild Steel Plates by X- and Gamma Ray

Sources and Some Observations on A.S.M.E., B.W.R.A., and D.I.N. Penetrameters. (*Brit. Weld. J.*, 1958, **5**, May, 244–247). Plates ¼, 1, 1½ and 2 in. thick were used. The D.I.N. wire penetrameter was shown to be the most reliable and X-rays were found superior to γ-rays though this advantage decreases as thickness increases.

Failures in Welding and Design. A. Mauchan. (*Engineer Foundryman*, 1958, **22**, April, 59–60; discussion 60–61). The failure of austenitic stainless steel road tankers was due to poor welding and severe restraint caused by poor design.

The Soundness of High-Temperature Brazed Joints in Heat-Resisting Alloys. A. Cibula. (*Brit. Weld. J.*, 1958, **5**, May, 185–201). Spreading temperatures and erosion properties are determined for Ni, Ag, Pd, Cu, Co and Mn brazing alloys on strips of austenitic stainless steel and Nimonic alloys. Fluxes were tested for high-temperature use. Conditions for high soundness of plane lap joints were found for each filler.

Brazing Alloys Tackle Heat Barrier. (*Steel*, 1958, **142**, May 19, 140–142). New high-temperature brazing alloys for joining heat-resistant alloys are reviewed.—A. G.

Brazing-Filler Metals Meet High-Temperature Needs. A. M. Setapen. (*Iron Age*, 1958, **181**, May 8, 110–111). Recent developments in types of brazing alloys for high temperature service are described. Lithium improves the fluidity and wettability of many standard alloys. For joining stainless steel, Inconel and other heat-resistant alloys a 70% Mn–30% Ni filler is being successfully used. Gold alloys are useful in cases where intergranular penetration cannot be allowed, and where high strength and good oxidation resistance are needed in the 1600° F region.

Combine Brazing and Hardening in One Operation. R. E. Wright. (*Iron Age*, 1958, **182**, July 17, 80–81). The General Electric Co.'s process for jet engine compressor blades is outlined.

Adhesive Bonding Methods. J. P. Wright. (*Metalw. Prod.*, 1958, **102**, Aug. 1, 1352–1358). A brief non-technical review of the use of bonding to replace welding, riveting, brazing and soldering. Operations are indicated in general terms and advantages are set out.

Effects of Oxidation on Adhesion of Polyethylene to Metals. F. J. Bockhoff, E. T. McDonel and J. E. Rutzler jun. (*Ind. Eng. Chem.*, 1958, **50**, June, 904–907). Stainless steels were used. Exposure of Agilene powder, scarfed film or commercial blown film to air, producing oxidation of the polymer, increases adhesion.

Room-Temperature-Cured Resorcinol Epoxide Adhesives for Metals. W. E. St. Clair and R. H. Moult. (*Ind. Eng. Chem.*, 1958, **50**, June, 908–911). Compositions for bonding of aircraft structures resistant to salt fog tests are developed.

Metal Surface Effects on Heat Resistance of Adhesive Bonds. J. M. Black and R. F. Blomquist. (*Ind. Eng. Chem.*, 1958, **50**, June, 918–921). Deterioration of bonds up to 550° F was studied. Iron and copper have adverse effects, the resistance of bonded Al being high.

Adhesion Using Molecular Models. D. Taylor jun. and J. E. Rutzler jun. (*Ind. Eng. Chem.*, 1958, **50**, June, 928–934). Studies of the adhesion of polythene and polyvinyl chloride to metals (Ti, Fe and stainless steel) by the use of models of the plastic on surfaces marked with the positions of the metal atoms are described.

Flame-Cutting of Air Hardening Steels. M. M. Griffith. (*International Acetylene Assoc. Proceedings*, 1956, 97–99). Procedure for ⅛ in. armour plate is given.

Oxy-Acetylene Cutting for Aircraft Jigs and Fixtures. B. Russell. (*International Acetylene Assoc. Proceedings*, 1956, 20–25). Machines used are briefly described.

Developments in Propane Torch Metal Cutting. J. R. M. Barclay. (*Engineer Foundryman*, 1958, **22**, April, 64–66). Details of the method and its advantages over oxy-acetylene methods are discussed.

MACHINING AND MACHINABILITY

Cemented Carbide Turning Tools. A. G. Gardner. (*Mech. World*, 1958, **138**, July, 308–313). A review of the material, tip mounting, tool design and setting, cutting speed, feed, depth of cut and coolants with further notes on grinding. Causes of tool failure are classified.

Alter Tool Structure for Longer Life. H. Chase. (*Iron Age*, 1958, **181**, Mar. 6, 117–119). The "Vitalizing" process developed by the Scandia Tool Co. for improving tool life is described. It was shown to improve rapidly as the amount of retained austenite decreased below 10%. "Vitalizing" is an electronic process employing a chemical bath. Ultra-high frequency current with varying resistances converts nearly all the austenite to martensite. In addition to high speed steels the process is beneficial to air-hardening steels, high-carbon high-chrome steels and oil-hardening steels. Application is to tools having sections of more than ½ in. in thickness, where there is liable to be retained austenite; sizes up to 20 in. dia. by 30 in. long can be treated. Tool life can be increased between 30 and 500%.—D. L. C. P.

Current Experimental Work with Sintered Oxide as a Cutting Tool. H. Eckersley and R. N. Cook. (*J. Inst. Prod. Eng.*, 1958, **37**, June, 341–362). A detailed review of recent publications on the use of Al_2O_3 tools. (16 references).

Carbide and Ceramic Tools in the Finish Turning of Steel. V. Šolaja. (*J. Inst. Prod. Eng.*, 1958, **37**, June, 363–372). A report on studies of tool wear and surface finish, with a review of previous work.

Metallurgical Factors and Machinability. J. Lomas. (*Machinery Lloyd*, 1958, **30**, June 28, 54–56). The definition of machinability and factors in its value depending on the tool and rate of working as well as on composition, microstructure and inclusions are discussed.

Testing of Machinability by Radioactive Methods. B. Colding. (*Acta Polytech., Mech. Eng. Series No.* 1, pp. 73.) Radioactivated cutting tools were used for the machining of 5 materials and wear was determined by chip radioactivity. ^{187}W was the isotope used, either β- or γ-rays being measured. After a brief running-in period the rate became constant till the tool was worn out.

Property of Free Machining. K. G. Lewis. (*Iron Steel*, 1958, **31**, Mar., 85–91; Apr., 133–140; May, 179–183). The author discusses the basic mechanics of the cutting process and of chip formation, dealing with the metallurgical factors, both physical and operational, which affect them. Additions of sulphur and lead to both ferrous and non-ferrous materials are considered, and the various views on the free-machining mechanism are reviewed. The influences of cutting fluids are discussed, and finally the effect of graphite in cast iron and of the cold-working of steel as free-machining aids are considered. (123 references).—G. F.

Suggestions on Designing For Machining Heat Resistant Alloys. R. D. Halverstadt. (*Steel Proc.*, 1958, **44**, Mar., 141–149, 164). The machining process is considered and various cases of drilling, tapping, turning and milling are discussed from the standpoint of design. Some stainless steels are included in the materials reviewed.

Studies on High Speed Machining. X. On the Machinability of Materials. H. Takeyama and E. Usui. (*J. Mech. Lab.*, 1958, **12**, May, 89–93). A machining formula based on the tool-chip contact area concept is tested and it is shown that the slope of the shear test curve near to rupture is closely related to contact area and so to machinability.

Machine Slashes Wheel Turning Time. (*Steel*, 1958, **142**, May 19, 138–139). A machine is described which processes wheels 33 to 42 in. in dia. in 2 min. 15 sec. Only the hub has to be bored before mounting on the axle.—A. G.

Surface Checks Aid Tool Control. (*Iron Age*, 1958, **182**, July 17, 79). Use of the Surfindicator for determining the point at which tool changing was needed is noted.

Effect of Subzero Treatment on Cutting Durability and Deflective Strength of High-Speed Steel. S. Koshiba and K. Tanaka. (*Nippon Kinzoku*, 1957, **21**, Apr., 225–227). [In Japanese]. Tests on low-tungsten, low-tungsten–cobalt, molybdenum–tungsten and 18–4–1 steel show that subzero treatment does not improve the cutting-durability of high-speed steel.—K. E. J.

Recent Investigations of the Trueing Process in Grinding. G. Pahlitzsch and W. Thöing. (*Werkstattstechnik Maschinenbau*, 1958, **48**, May, 254–259). The importance of the shape of the grinding disc, and the influence on shape, as measured by "effective depth of roughness" of the trueing operation are emphasized. Various factors affecting this operation are measured and discussed.

Abrasive Belts Cut Roll Grinding Costs. (*Steel*, 1958, **142**, Mar., 31, 92–94). The use of power-driven abrasive belts for grinding operations is described. Lathes or roll-grinding machines can be converted for belt grinding. Abrasive costs may be higher than with wheels, but the high cutting rate and ease of abrasive-size changes give advantages in time and money.—D. L. C. P.

Punched-Tape Units Control New Type Transfer Line. (*Iron Age*, 1958, **181**, Mar. 20, 106–108). An electronically-controlled machine tool line, specially designed for small and medium quantity production runs, is described. Three units carry out milling, drilling and boring operations and work is positioned at each by a transfer bar. Control is by punched-tape units.—D. L. C. P.

On the Finishing of Blade Edges by Electrolytic Polishing. III. T. Kawakita. (*Nippon Kinzoku*, 1957, **21**, Feb., 96–99). [In Japanese]. Rough-honed blade edges were finished by electrolytic polishing using H_3PO_4–CrO_3 solution. Optimum conditions were established.

CLEANING AND PICKLING

6 Ways to Safeguard Good Appearance, Long Life [In Cleaning Stainless Steel]. (*Ind. Fin.*, 1958, **10**, July, 28–29). Brief notes with illustrations.

Metal Polishing Developments. D. J. Fishlock. (*Mech. World*, 1958, **138**, Mar., 102–106). A review contrasting new methods with old. Selection of technique is considered. Various new devices, especially automatic machines, are noted and plating developments are referred to.

Developments in Chemical Cleaning. D. J. Fishlock. (*Electropl. Met. Fin.*, 1958, **11**, July, 239–242). A discussion of the properties of the Grisiron range of proprietary cleaners is given.

Chemical Polishing (II). L. F. Spencer. (*Met. Fin.*, 1958, **56**, April, 62–67). Applications of chemical polishing to ferrous and non-ferrous metals and costs of the process are considered. (39 references).

A Glossary of Barrel Finishing Terms. (*Prec. Met. Mold.*, 1958, **14**, July, 43–44).

Removal of Scale from the Surface of Rolled Sheet by the Steam-Gas Explosive Method. M. I. Bazhenov-Korchemnyi and L. I. Danilov. (*Metallurg*, 1958, (6), 27–30). Since both the birch twig and hydraulic methods were not completely satisfactory, trials were carried out with slotted rolls. Cooling water entered the slots and formed high pressure steam which exploded out of the slots and blew off the scale.—B. S.

Flame Cleaning. J. J. Harrison-Smith. (*Engineer Foundryman*, 1958, **22**, April, 68–69; discussion, 69–70). Methods and advantages of removal of scale and rust by oxy-acetylene heating before painting are described.

Surface Treatment of Ingots and Semi-finished Steel by Flame Cleaning. H. E. Habbig and A. Pfeuffer. (*Schweissen, Schneiden* 1958, **10**, April, 135–137). The economy and usefulness of flame cleaning equipment in steelworks is discussed.—U. E.

Mechanical Scarfing. J. H. Such. (*Inter-*

national Acetylene Assoc. Proceedings, 1956, 116–125). Use of oxygen and oxy-acetylene jets for conditioning steel ingot surfaces before rolling is reviewed, mostly in the U.S.

How to Treat Pickle Liquors in a Small Plant. E. W. Lang. (*Met. Prog.*, 1958, **73**, May, 93–96). The operation and control of plant for batch crystallization of copperas from liquor containing 25% $FeSO_4$ and the return of the regenerated liquor containing 15% $FeSO_4$ are described.

Evaluation of the Blaw-Knox–Ruthner Pilot Plant Programme. J. H. Strassburger. (*Blast Furn. Steel Plant*, 1958, **46**, June, 587–594). The $FeSO_4$ which is formed in the pickler is concentrated and transferred to the reactors where it is contacted with HCl. The reaction taking place is: $FeSO_4 + 2HCl = FeCl_2 + H_2SO_4$. The sulphate is recovered as sulphuric acid. The chloride is roasted in furnaces. The reaction taking place here is: $FeCl_2 + H_2O = FeO + 2HCl$. The HCl is also recovered. Only the unavoidable losses of both acids have to be replaced. The FeO can be sintered and charged to the blast furnace. The plant is described in detail and the economics of the process evaluated.

PROTECTIVE COATINGS

Electrodeposited Metallic Coatings. A.S.T.M. Committee B-8. (*A.S.T.M. preprint*, 1958, No. 14, pp. 15). **Summary of Observations of Factors Influencing Adhesion of Organic Coatings to Chromium Plated Surfaces.** (5–6). **Cleaning Metals Prior to Electroplating.** (7–15).

Electroplating Installation at Vauxhall Motors. (*Ind. Fin.*, 1958, **10**, June, 32–35). Two automatic plants for the deposition of Cu, Ni and Cr are described.

5 Rules to Follow in Designing for Plating. C. F. Nixon, D. M. Bigge and W. L. Pinner. (*Mat. Design Eng.*, 1958, **47**, June, 118–120). A chapter from the book edited by A. K. Graham, Reinhold Publishing Co., 1955.

Barrel Plating is Finding its Place. D. J. Fishlock. (*Metalw. Prod.*, 1958, **102**, June 27, 1115–1121). A review of apparatus and processes.

Some Notes on American Experience in Plating with Ultrasound. H. A. Reich. (*Met. Fin. J.*, 1958, **4**, Mar., 97–98). The author discusses the uses of ultrasound in plating, considering both the development of many unprecedented finishing processes and also the improvement of standard procedures.

Chromium Plating Baths Containing Sodium Fluosilicate. III. Operating Conditions. T. Hayashi, S. Miwa and S. Takagi. (*Rep. Gov. Ind. Res. Inst.*, 1956, **5**, Aug., 373–377). [In Japanese]. A bath with 250 g/l CrO_3, 5g/l Na_2SiF_6 and 1·5 g/l H_2SO_4 was examined at 40°, 50° and 60° C. Temperature had little effect on cathode efficiency at c.d. > 40 A/dm², (25–26%). Covering power at 55° C was almost independent of Cu, Zn and Fe but current efficiency falls sharply over 25 g/l Cu, 25 g/l Zn or 17 g/l Fe.

Cold Chromium Plating in Russia. A. J. Steiger. (*Met. Fin.*, 1958, **56**, April 56–57). Details of the use of a solution containing H_2CrO_4 250, NH_4F 4 gm/l and operated at 15–25° C at a current density of 8–12 amp/dm² for hard chroming of punches are given.

Chromium Plating Thickness. J. B. Mohler. (*Met. Fin.*, 1958, **56**, April, 58, 67). Data on thicknesses of Cr plating used for decorative and engineering purposes are given.

Stress-Free Nickel Plating. (*Met. Prog.*, 1958, **73**, April, 90–92). A bath containing nickel sulphamate 60, boric acid 4 and anti-pitting agent 0·05 oz./gal gives deposits which are either stress-free or in a state of compressive stress. Applications of these coatings are described.

The Anodic Behaviour of Various Grades of Nickel in Electrolytic Nickel Baths (I). E. Raub and A. Disam. (*Metalloberfläche*, 1958, **12**, June, 161–167). Five different types of nickel anode were examined microscopically, and anode potential-current density curves established in several baths at different temperatures and pH values.—L. D. H.

The Structure of Nickel Electrodeposits in Relation to Some Physical Properties. D. J. Evans. (*Trans. Farad. Soc.*, 1958, **54**, July, 1086–1091). Studies on preferred orientation, internal stress, grain size, hardness and brightness are reported. The orientation is incidental and not a cause of brightness.

Multilayer Nickel Coatings. (*Met. Prog.*, 1958, **73**, June, 95–96). The corrosion resistance of a given thickness of electrodeposited nickel is improved by using a semi-bright undercoat (80% of total thickness) and the remainder of fully bright levelling nickel.

The Protective Value of Tin–Nickel Alloy Deposits on Steel. F. A. Lowenheim, W. W. Sellers and F. X. Carlin. (*J. Electrochem. Soc.*, 1958, **105**, June, 338–346). Outdoor corrosion tests show that tin–nickel deposits provide corrosion resistance which is equal to that of nickel-chromium deposits in marine atmospheres and better in industrial atmospheres.

Quality of Galvanising Zincs. A. Herz. (*Mét. Constr. Mécan.*, 1958, **90**, May, 375–381). The properties of pure zinc and of various alloys are discussed from the aspect of the quality of zinc to be used for galvanizing.

Experiences in Electroplating and Related Processes. J. Chadwick. (*Electropl. Met. Fin.*, 1958, **11**, July, 225–228). Cd and Zn plating experiences are included.

Continuous Galvanising by the Armco-Sendzimir Process. I. C. Gwynne. (*Sheet Metal Ind.*, 1958, **35**, July 497–506, 534). An illustrated account is given of the Ebbw Vale line.

Continuous Galvanising Line. (*Elect. Rev.*, 1958, **163**, July 18, 95–98). An account of the electrical equipment of the Richard Thomas and Baldwins line.

Britain's Largest Galvanising Line is Gas-Fired. (*Ind. Gas.*, 1958, **21**, July, 24–25). An account of the Ebbw Vale installation with special reference to the heating arrangements.

Continuous Galvanizing Line at R.T.B.'s Ebbw Vale Works. Production of Speltafast. (*Iron Coal Trades Rev.*, 1958, **177**, July 25, 206–207). Speltafast is a sheet with a Zn coating capable of standing up to forming operations.

Conveyorised Systems for Galvanising. (*Ind. Fin.*, 1958, **10**, July, 31, 33). The Crittal Manufacturing Co. Braintree Works hot dip process installation is briefly described.

Modern Tinplate Production. (*Met. Bull.*, 1958, Apr., Special Issue, 32, 34, 35, 37). An illustrated account of the Steel Co. of Wales mills.

The Technology of Tinplate Manufacture: A Generation of Progress. W. E. Hoare. (*Met. Bull.*, 1958, April, Special Issue, 12–24). An illustrated review.

On the Oxidation Resistance of Aluminium-coated Cast Iron (1st Report). T. Saga, O. Miyakawa and K. Kôda. (*Nippon Kinzoku*, 1957, **21**, Sept., 548–551). From an investigation of hot-dip aluminized grey and white irons, it is concluded that the coating markedly reduces weight increase and dimensional growth of grey cast-iron, caused by oxidation; the thickness of the coating has no marked effect. In aluminized white irons the weight increase is no more than 10% but dimensional changes are hardly altered by aluminizing.

Study on Aluminium-coated Steel. VI. On the Formation of the Interfacial Alloy Layer. T. Saga and K. Sato. (*Nippon Kinzoku*, 1957, **21**, Feb., 110–113). [In Japanese]. Results are given for sheet of thicknesses in the range 0·2–1·2 mm coated by the single- and double-bath methods. The formation of the alloy layer is affected by sheet thickness, being mainly influenced by the local temp. drop near the base metal. Cold working does not affect it: internal strain is removed during dipping. The alloy layer is thickened by preheating, whether as a result of the double bath or otherwise. (10 references). **VII. Influence of the Chemical Composition of Steel on the Formation of Interfacial Alloy Layer.** T. Saga and M. Yasui. (114–117). Si, Cr or Mn in steel act as inhibitors in the same way as when added to the bath, but Si is more effective in the bath. The softening effect of Si increases with its content. The mean chemical composition of the alloy layer produced with pure Fe and Al is somewhat higher in Fe content than $FeAl_3$.—K. E. J.

Automation of a Barrel Plating Line. (*Met. Fin. J.*, 1958, **4**, Mar., 75–79). Details are given of the design and layout of a new automatic independent-cycling Cd barrel-plating line, installed at the Detroit works of Bellevue Plating Co. It incorporates the "Dial-A-Cycle" principle of completely automated cycling for each barrel, independent of all others.—G. F.

Experimental Consideration of a New Process of Metallic Protection. M. Serra. (*Rev. Ciencia Apl.*, 1958, **12**, May-June, 222–237). [In Spanish]. Details of the design, construction and experimental use of a semi-industrial plant for the continuous coating of iron wires with aluminium by molten bath immersion are given. The results indicate that for each molten bath composition there is a critical temperature for coat formation and this temperature coincides with that giving a minimum thickness of coat. The best results were obtained with a bath containing 94·7% Al, 5% Si, 0·14% Cr, 0·14% Mo and 0·005% Be. The main purpose of the Be addition is for brightening the coat, it is not necessary, like the Cr and Mo, for improving the mechanical properties. The critical temperature for the bath was found to be 770° C, or 740° C in the absence of Be. Photographs of the equipment and the microstructures produced are given.—F. S.

Stainless Gets New Sales Appeal. (*Steel*, 1958, **142**, May 12, 88–89). The Permyron process for colouring metal is mentioned, which permits severe forming after colouring. Combined with the method of rolled-in patterns, this is stated to have many applications.—A. G.

Plastic Coated Steel. (*Iron Steel*, 1958, **31**, Mar., 106). A brief account is given of the nature and properties of "Stelvetite," the new P.V.C.-coated steel made by John Summers & Sons Ltd.—G. F.

Protecting Metals at High Temperatures. A. F. Hofstatter, (*Mat. Design Eng.*, 1958, **47**, April, 115–119). The properties and applications of ceramic and metallic coatings and silicone and sodium silicate paints are discussed.

Continuous Pretreatment Processes with Particular Reference to "Nitec" Sheet for Vitreous Enamelling. W. H. F. Tickle and D. A. Winton. (*Inst. Vit. Enamel*, 1958, **8**, May, 81–84). A brief account of pre-coated sheet, actually a commentary on a film.

Porcelain Enamel—Corrosion Protection for Chemical Process Equipment. (*Ind. Eng. Chem.*, 1958, **50**, June, 75A–76A). A brief review of properties.

The Application of Vitreous Enamel by the Electrostatic Process. S. Hallsworth. (*Inst. Vit. Enamel*, 1958, **8**, May, 75–80).

Continuous Electric Enamelling Furnace. (*Process Control Autom.*, 1958, **5**, July, 273). An installation built by Ferro Enamels Ltd. and installed at Belling & Co. Ltd. works, Enfield, is described.

Glass Coatings—A Good Way to Protect Metals. C. E. Bullock and F. Nelson. (*Mat. Design Eng.*, 1958, **47**, Mar., 106–109). New types of glass with unusually favourable combinations of properties are leading to new uses for glass coatings for the protection of metals. The authors describe the chemical and mechanical properties of the coatings and their resulting performance, and mention design factors which must be considered.

Frit for Vitreous Enamelling. (*Engineer*, 1958, **206**, July 25, 139–140). Procedure at Ferro Enamels Ltd. is outlined.

Self-Opacifying Titania Enamels for Cast Iron. (*Ind. Fin.*, 1958, **10**, June, 44). The properties of TiO_2 enamels suitable either for direct application or as annealing-type drip coats are given. High covering power and improved durability are claimed.

Quantitative High-Temperature Oxidation of Porcelain Enamelled Iron. H. G. Lefort and A. L. Friedberg. (*J. Amer. Ceram. Soc.*,

1958, **41**, June, 216–226). Bare and enamelled iron samples were heated in air for long periods and increase in weight was observed. Thickness of enamel, thickness of nickel plating and metal surface preparation were varied. After an initial linear period the rate on all specimens became parabolic. With rise of temperature up to 1850° F the advantage of enamel, and the smaller advantage of Ni, decreased.

Problems of Simultaneous Heat-Hardening and Ceramic Coating of No. 420 Stainless Steel. E. L. Bradley. (*Amer. Ceram. Soc. Bull.*, 1958, May, 222–226). The cooling contraction curve is shown to be useful for determining the optimum softening point of the enamel.

The Prevention of Corrosion by Paints Containing Metallic Pigments or Buffer Substances. C. E. Bird. (*Abstracts of Dissertations, Cambridge*, 1954–1955, 1957, 180–182). The Fe–Al system is examined. Sprayed coatings of Al and Zn on steel, Al and Fe paints on steel and the effects of ferrous ions and rust on the corrosion of Al are studied.

I.C.I. Marine Research Station. (*Chem. Ind.*, 1958, July 26, 947–948). A description of the new station for corrosion studies and marine paint exposures.

Epoxide Resins in Anti-Corrosion Finishes. F. L. Farr. (*Indust. Finish*, 1957, **9**, Dec., 986–991). Industrial and marine applications of epoxide resin paints are given.

CLAD SHEET AND HARD-FACING

Stainless Clad Plate: Don't Look at Cost Alone. D. T. Smith. (*Iron Age*, 1958, **181**, Apr. 17, 112–113). The technical advantages as opposed to economic advantages of stainless clad plate are considered. Useful properties are–high ductility coupled with corrosion resistance; high thermal conductivity; ease of welding, burning, machining, shearing and forming.—D. L. C. P.

Cladded Steels. (*Mét. Constr. Mécan.*, 1958, **90**, May, 371–374). Reasons are given for the rapid increase in the use of composite steel sheets. Methods used by the Compagnie des Ateliers et Forges de la Loire for the fabrication and testing of such plates are described.—A. G.

A Survey of Factors Controlling Metallic Diffusion from the Gas Phase. R. L. Samuel. (*Murex Rev.*, 1958, **1**, (18), 501–527). A comprehensive review of the mechanism of coating by diffusion. Chromizing and siliconizing of iron and other metals are considered. (34 references).

The Formation of Metal Silicides with Silicon Tetrachloride (1st Report). Siliconising of Steel. M. Someno. (*Nippon Kinzoku*, 1957, **21**, Aug., 490–494). Investigations are reported of steel siliconized with a $SiCl_4$ atmosphere. The effects of gas flow rate, temperature and other variables differed according to whether the atmosphere was reducing or neutral. There was a discontinuity in the reaction rate of siliconizing near 900° C.

Process Upgrades Steel Parts. (*Steel*, 1958, **142**, June 9, 90–91). Properties of chromized parts and applications of the chromizing process are briefly discussed.—A. G.

Chromised Steel of Higher Corrosion Resistance. N. F. Vyaznikov and A. N. Popandopulo. (*Metallov Obrabotka Met.*, 1958, (3), 60–61). Additional carburizing improves hardness and resistance to pitting.

Hard-Facing of Oil Tools. J. S. Goodwin. (*International Acetylene Assoc. Proceedings*, 1956, 33–37). WC particle application is outlined.

Develop New Method for Bonding Aluminium to Iron. (*Light Metal Age*, 1958, Apr., 18). The B.M.I. process is briefly noted in which Al in a die-casting machine is used to form the bond for brake drums.

POWDER METALLURGY

Metal Powders and Metal Powder Products. A.S.T.M. Committee B-9. (*A.S.T.M. preprint*, 1958, No. 15, pp. 16). **Density and Interconnected Porosity of Sintered Powder Metal Structural Parts and Oil Impregnated Bearings** (7–8). **Average Particle Size of Refractory Metals and Compounds by Fisher Subsieve Sizer** (9–12). **Apparent Density of Refractory Metals and Compounds by Scott Volumeter** (13–14). **Determination of the Compressibility of Metal Powders** (15–16). (All tentative recommended methods).

When You Can—Design for Sinterings. (*Prec. Met. Mold.*, 1958, **16**, July, 32). Advantages in the construction of moving parts for a power mower are set out

Sintered Iron-Base Alloys for Machine Parts. I. Properties of Electrolytic Iron Powder, Reduced Iron Powder and Their Sintered Compacts. E. Hayashi, G. Matsumura, T. Watanabe, T. Takeuchi and Y. Minamitani. (*Rep. Gov. Ind. Res. Inst.*, 1956, **5**, Aug., 361–368). [In Japanese]. Electrolytic iron powder had higher apparent densities than reduced powder but inferior flow rate. Reduced is superior to electrolytic in compactability. Apparent density, linear shrinkage, hardness and tensile strength increase with sintering temperature except beyond the A_3 point, and they increase again above 1000° C. The mechanical properties of sintered reduced are superior to those of electrolytic though the difference becomes small above 1200° C.

What to Consider in Choosing Powder Metallurgy. K. H. Roll. (*Mat. Design Eng.*, 1958, **47**, April, 106–109). Properties and costs of components made by powder metallurgical methods are compared with other forming processes.

Powder Parts Made Without a Press. (*Steel*, 1958, **142**, May 19, 162–163). The application of slip casting to the fabrication of refractory metal parts is described.—A. G.

Sintered Iron-Base Alloys for Machine Parts (VI). Gas Carburization of Sintered Iron-Base Compacts (I). Y. Fujii, E. Hayashi, T. Watanabe and M. Yasuda. (*Rep. Gov. Ind. Res. Inst.*, 1958, **7**, May, 329–336). [In Japanese]. Charcoal-converted town's gas was used and the thickness of the case was found to depend upon pressure, temperature and time, where the compacting pressure was the most important factor. Pressures of > 7 t/cm², temperatures ~ 830° and times of 30–60 min. are recommended.

FERRITES, CERMETS AND CARBIDES

High-Frequency Magnetic Losses in Ferrites. K. Ohta and S. Chikazumi. (*Metal Physics*, 1957, **3**, July, 129–140). [In Japanese].

Stability of Ferrites Under High Pressure. N. Kawai and S. Kume. (*Metal Physics*, 1957, **3**, July, 155–156). [In Japanese].

On the Preparation of Single Crystals of Nickel Ferrite. I. Kushima, T. Amanuma and K. Tanaka. (*Metal Physics*, 1957, **3**, May, 115–116). [In Japanese].

Study of the System Fe_2O_3, MnO–Ga_2O_3,MnO. M. Lensen and A. Michel. (*Compt. Rend.*, 1958, May 12, 2766–2769). The continuous variations in parameters, Curie point and magnetic moment indicate total solid solution between the ferrite and gallate of Mn. Deductions are made about lattice structure.—A. G.

On Spinel-Type Lithium Ferrite. T. Takei and S. Okamoto. (*Rep. Sci. Res. Inst.*, 1958, **34**, Mar., 125–128). [In Japanese].

Above 2500° F What Material to Use? L. D. Loch. (*Chem. Eng.*, 1958, **65**, June 30, 105–109). Carbides and cermets are reviewed.

Process of Sintering in $FeAl_2$–Ni–Fe System. M. Sugiyama, Y. Mishima and K. Shida. (*Ann. Rep. Eng. Res. Inst. Tokyo Univ.*, 1957, **16**, Sept., 35–40). [In Japanese]. Density, resistivity, microstructure, X-ray diffraction patterns and other properties were followed during sintering of the system containing Al 12, Ni 25 and Fe 63%. At a sintering temp. of 700° C, changes caused by diffusion were noted; at 1200° C a partial cast structure was found; and at 1250° C rapid alloying of particles and appearance of σ-phase were observed.—K. E. J.

PROPERTIES AND TESTS

Arc Image Furnace Produces Temperatures Above 7000° F. (*Ind. Heat.*, 1958, **25**, Apr., 728–730, 734). A carbon arc is focused by parabolic mirrors to produce a source of heat for high temperature research studies.

On the Relationship between Poisson's Ratio and the Plastic Properties of a Material. I. I. Cherkasov. (*Zhur. Tekhn. Fiz.*, 1952, **2**, 1834–1837, Translation 19RL–T/C83). The equation is tested for normalized and hardened steels and for other metals.

Internal Friction Measurements. A. Sosin, L. Bienvenue and H. Schlein. (*Rev. Sci. Inst.*, 1958, **29**, July, 657–659). An electrical modulus decrement measuring device is described. An initial steady oscillation is set up and its fading measured. Most difficulties have arisen in setting up the initial steady state.

Apparatus for Measuring the Internal Friction of Rigid Test Pieces. G. Collette. (*Compt. Rend.*, 1958, May 12, 2756–2758). The apparatus described is for determining the internal friction of plate test pieces at moderate temperatures using low-frequency alternating stresses. It has been used to measure the limit of solubility of N_2 in ferrite at 100° C.—A. G.

Further Studies of Properties of Stresscoat Under Dynamic Loading. J. W. Dalley, A. J. Durelli and V. J. Parks. (*Proc. Soc. Exper. Stress Anal.*, 1957, **15**, (2), 57–66). Dynamic and static strain sensitivities as a function of coating number, curing temperature and thickness were examined with durations of maximum load from 8–500 millisec. Strain sensitivity is nearly a log. function of loading time and the largest influence is exerted in periods below 50 millisec.

A Dynamic Strain Calibration Device. S. Okubo, A. J. Durelli and J. W. Dally. (*Proc. Soc. Exper. Stress Anal.*, 1957, **15**, (2), 67–72). A cantilever beam calibrated device for loading and timing is described. About 8 millisec. are required to impose a maximum strain of 1200 microinches/in. for times varying from 7–430 millisec.

Experimental Solution of Stress Diffusion Problems. C. Riparbelli. (*Proc. Soc. Exper. Stress Anal.*, 1957, **15**, (2), 73–84). Stresscoat is used on the structure or on a model.

A Simplified Method of Obtaining Principal Stress Information from Strain Gauge Rosettes. P. K. Stein. (*Proc. Soc. Exper. Stress Anal.*, 1957, **15**, (2), 21–38). Mathematical. *Some* **Simplications in Rosette Analysis.** W. M. Murray. (39–52).

A Rapid Method of Waterproofing Bonded Wire Strain Gages. F. E. Wells. (*Proc. Soc. Exper. Stress Anal.*, 1957, **15**, (2), 107–110). The gauge is covered with a soft wax, Al foil, and an epoxy resin.

Internal Friction of Polycrystalline Iron due to Point Imperfections. R. R. Hasiguti, G. Kamoshita and N. Igata. (*Metal Physics*, 1956, **2**, July, 163–164). [In Japanese].

Causes and Prevention of Surface Fissures on Hot-Worked Steel Products (1st Report). Influences of Heating Conditions, Scaling Loss and Tramp Elements on Surface Fissures. T. Ohtake, K. Ishizaki and N. Eguchi. (*Nippon Kinzoku*, 1957, **21**, Sept., 536–540). From hot bending tests it was concluded that surface fissures occur after a critical amount of scaling and above this increase in depth with increased scaling. The threshold value of scaling increases with heating temperature and the presence of tramp elements, and is affected by the heating atmosphere.

An Introduction to Metallurgical Thermodynamics. J. N. Pratt. (*J. B'ham Met. Soc.*, 1958, **38**, June, 56–75).

On the Electronic Structure of Transition Metals and Alloys and of Heavy Metals. J. Friedel. (*J. Phys. Radium*, 1958, **19**, June, 573–581). Fe is included, discussed as an impurity in Cu, Al etc.

Drill Steel Developments in the Laboratory. F. R. Anderson. (*Bull. School Mines Met. Univ. Missouri*, 1958, (95), 31–37). A contribution to a symposium on mining research. Measurements of tensile and other stresses in rock boring tools are examined.

Carbon-Manganese Steel: Yield Point and High-Temperature Proof Stress. J. Glen. (*Iron Steel*, 1958, **31**, May, 165–171). The author records experimental work carried out to determine the proof stress values at temperatures up to 450° C for C–Mn steels falling within the range 24–35 t/in.² ultimate

tensile strength at atmospheric temperature. The samples were obtained from commercially treated plate, and covered most of the C–Mn steels used for pressure vessels and made by the basic O.H. process. They included balanced Si-killed, and Al-killed steels in the as-rolled, normalized, and stress-relieved conditions.

Why Threaded Test Bars Lower A-286 Stress-Rupture Values. H. T. McHenry and R. K. Pitler. (*Iron Age*, 1958, **181**, June 12, 87–89). Bolts and threaded test specimens of A–286 austenitic heat-resistant steel failed to meet stress-rupture specifications. This is attributed to the geometry of the test piece rather than to the material itself.—A. G.

Special Bending Test for Electrodeposited Coatings. J. Edwards. (*Trans. Inst. Met. Finishing*, 1958, **35**, Summer, 101–106). A former in the shape of a log. spiral is used and the radius of curvature where cracks first form is read off from graduations.

On the Bending Test, Sharpness, Finishability and Durability of Thin Blade Cutlery. II. T. Kawakita. (*Nippon Kinzoku*, 1957, **21**, Feb., 92–96). [In Japanese]. Carbon steel tape 0·13 mm thick was quenched on a water-cooled plate under various conditions of drawing speed, quenching temp. and tempering temp.; razor-blades were made from the samples. Tests showed that samples of almost equal hardness had greatly differing bending behaviour; there is an optimum value of bending for good finishing qualities.—K. E. J.

On the Mechanical Properties of Rapid-Heated and Quenched Steel. IV. S. Owaku and K. Iijima. (*Nippon Kinzoku*, 1957, **21**, Apr., 206–210). [In Japanese]. Results are given for 0·81% C steel wire. Semi-rapid heating and quenching produce similar values for hardness and for fatigue limit when the tempering temp. exceeds 300° C. With lower tempering temp. superior values for fatigue limit and bend-breaking stress are shown by semi-rapid heated and quenched specimens. These results are explained in terms of micro-stresses induced by martensite formation in the grains.—K. E. J.

Torsion of Cylindrical and Prismatic Bars in the Presence of Steady Creep. S. A. Patel, B. Venkatraman and P. G. Hodge jun. (*J. Appl. Mech.*, 1958, **25**, June, 214–218). Mathematical.

Behaviour of Thick-Walled Steel Cylinders Subjected to Internal Pressure. B. Crossland and J. A. Bones. (*Inst. Mech. Eng. Advance Copy*, 1958, pp. 20). Tests on various steels are described and yield pressure, collapse pressure and ultimate pressure are related to shear stress–strain data.

On Breaking Behaviour in Charpy Impact Bending Tests. M. Tanaka and S. Umekawa. (*Nippon Kinzoku*, 1957, **21**, Apr., 221–225). [In Japanese]. Load-time curves in the Charpy impact bending test were obtained using a force detector employing a crystal of Rochelle salt. Tests on structural steel, tool steel and pure copper show that there are four types of breaking behaviour, which differ even when the Charpy values are the same. The tests have been used to show how the cold-notch shortness of hot-rolled plate can be improved.—K. E. J.

Comparison of Charpy Vee, Charpy U and Izod Notched Bar Tests on Steel over Transition Temperature Range. A. W. Johnston. (*D.S.I.R. Mech. Eng. Res. Lab. PM 247 AB Div. No.* 9/58, 1958, April, pp. 81). Three mild steels and three alloy steels with 2·5% Ni–Cr, 3·5% Ni–Cr and 0·4% C, 1% Ni were tested over the range − 196° C to + 110° C with both 10 × 10 mm and 10 × 5 mm test pieces. No simple relationships were found, but various limited correlations applying to a few of the tests are pointed out. Energy absorbed in the tests was determined.

Fracturing Under Impulsive Loading. J. S. Rinehart. (*Bull. School Mines Met. Univ. Missouri*, 1958, (95), 46–84). Studies on the effects produced by firing explosives in thick cylindrical metal specimens are reported. Fracture patterns are examined and shown to be non-random, the stress inhomogeneities leading to fracture being qualitatively established.

Variation in Fatigue Properties over Individual Casts of Steel. Part II: Investigation of a Cast of Steel to B.S.S. 970 En 24. E. Ineson, J. Clayton-Cave, R. J. Taylor. (*J.I.S.I.*, 1958, **190**, Nov., 277–283.) [This issue.]

X-Ray Determination of Heterodiffusion Coefficients in Alloys with Components Differing Considerably in X-Ray Absorption. B. Ya. Pines and I. V. Smushkov. (*Zhur. Tekh. Fiz.*, 1958, **28**, (3), 661–667). Examination of a method of calculating heterodiffusion coefficients from radiographs of specimens of binary alloys is reported.

Low-Cost Cryogenic Steel Goes Commercial. J. Procter. (*Chem. Eng.*, 1958, 65, July 14, 160, 162, 164). A 9% Ni steel with high strength and impact resistance at − 320° F is described. Difficulties in forming are noted, possibly due to the coefficient of expansion, and these can be avoided by modifying the time-temperature cycle.

METALLOGRAPHY

Metallography of Low-Carbon Low-Alloy Steels. P. P. G. L. Siriwardene. (*Abstracts of Dissertations*, Cambridge 1954–1955, 1957, 194–195). Studies on the improvement of Mo-containing low-carbon steels by B are reported. Optical and electron microscopy was applied to hardness studies and ferrite grain size refinement appeared. Carbides were isolated and examined.

Galvanic Corrosion. J. Bonaston. (*Inspection Eng.*, 1958, **22**, May-Jun., 50–56). The mechanisms of corrosion due to dissimilar metals, crevices and stray currents are discussed with examples from aircraft practice. The uses of steel rivets for Al and Al rivets for steel are considered.

Corrosion Research and its Industrial Background. W. H. J. Vernon. (*Chem. Ind.*, 1958, June 28, 818). A report of the Spring Lecture.

Corrosion in the Programme of Investigations by Cebedeau. E. Leclerc. (*Corros. et Anticorros.*, 1957, **5**, June, 185–188). An outline is given of research actually being carried out, or contemplated, at the Centre belge d'Etude et de Documentation des Eaux.—L. D. H.

A New Method of Estimating the Cost to the Nation of Corrosion. F. Podbrežnik. (*Zaštita Mat.*, 1956, **4**, Jan., 4–11, 28).

Corrosion Experiences with Dissimilar Metals. A. A. Brouwer. (*Ind. Eng. Chem.*, 1958, **50**, Apr., 73A–74A). 8 case histories are briefly given.

A New Aspect of the Corrosion of Acid-Resisting Austenitic Steels and Welding Seams. (Transcrystalline Corrosion of Chromium–Nickel–Molybdenum—Copper Steel Type Kh23N23M3D3, Caused by Compression Deformation). B. I. Medovar, N. A. Langer and Yu. V. Latysh. (*Fiz. Met. Metalloved.*, 1957, **5**, (1), 184–186). Cr–Ni stainless steels and acid-resisting austenitic steels not containing Ti and Nb are subject to intercrystalline corrosion because of the separation of Cr along the grain boundaries and the impoverishment of this element in the peripheral regions of the austenitic grains. Experimental results have shown that when subjected to boiling in acid of various concentrations, after cold hardening and torsion stresses, the surface of the specimens previously so treated, show coarse corrosion fissures, whilst specimens subjected to a weaker cold hardening or weaker torsional stresses show surfaces with much finer defects. The corrosion destruction of welded joints has a transcrystalline character both in the base metal and in the metal of the weld.—L. H.

Corrosion of Cutlery. E. A. Oldfield and D. Sheppard. (*Corrosion Tech.*, 1958, **5**, June, 187–189). Tests on stainless steel in sea water and salt solutions are reported and corrosion in service, packing and protection by inhibitors are also reviewed. Recommendations are added.

Corrosion of Ten Metals in Boiling Hydrochloric Acid when in Contact with Rhodium, Palladium, Iridium and Platinum. W. R. Buck and H. Leidheiser, jun. (*Nature*, 1958, **181**, June 14, 1681–1682). Iron was included in the study in 2N HCl. In contrast to other metals the rate increased only 1·25–1·42 times in contact with the noble metals.

Corrosion-Resistance of Spheroidal-Graphite Iron. F. L. LaQue. (*Found. Trade J.*, 1958, **104**, June 12, 711–714). Acid corrosion and performance in marine, atmospheric and underground environments are reviewed. The usefulness of the materials in tankers is underlined.

Corrosion in Stainless Steels, Passivation. Bertolini. (*Ing. e Indust.*, 1957, **25**, Oct., 116–120). [In Spanish]. The various types of possible corrosion attack are listed, and intergranular corrosion is dealt with more fully. Brief details of the mechanism of passivation are given.—P. S.

Corrosion of Steel in Moist Air. U. R. Evans. (*Chem. Ind.*, 1958, June 7, 681). A letter. A difference in the corroding effect of large and small water drops in a closed vessel is confirmed. It is attributed to the probability of the area under the drop including a centre sensitive to dissolved oxygen.

Influence of Nickel on Inter-granular Corrosion of 18% Chromium Steels. J. R. Upp, F. H. Beck and M. G. Fontana. (*Trans. A. S. M.*, 1957, **50**, preprint 51, pp. 18). It was found that 18/8 stainless steel was susceptible and that 18% Cr steel was resistant to intergranular corrosion when heated to 650–760° C. Alloys containing intermediate quantities of nickel showed that the transition occurs at about 2·5 to 3% Ni; steels with a lower Ni content should be heat treated like ferritic steels and those with a higher content like austenitic steels. All alloys showed good resistance after water quenching from 760° C.—E. E. W.

Corrosion of Metals in Ethylene Glycol Solutions. R. J. Agnew, J. K. Truitt and W. D. Robertson. (*Ind. Eng. Chem.*, 1958, **50**, Apr., 649–656). Cast iron, steel and other metals used in internal combustion engines are studied.

White Rust of Galvanised Articles—Its Origin and Prevention. K. Daeves. (*Draht*, 1957, **8**, Aug., 334–335; *Wire*, 1958, Apr., 22–23). Moisture and condensation are pinpointed as causes of white rust, and measures should be taken to prevent them by ventilation. Reference is also made to dipping in a weak acid chromate solution for protective purposes.—J. O. W.

Detection and Rapid Localisation of the Corrosion Resistance of a Protective Coating on Metallic Substances. G. Blet. (*Peint. Pig. Ver.*, 1958, **34**, Apr., 160–163). A new electrical method of determining the efficacy of non-conducting coatings on metal surfaces in a few minutes is described. The method measures the porosity of the coatings and the results compare very well with those of classical methods of testing in boiling salt solutions.—B. G. B.

A Study of the Compatibility of Some Creep-resistant Steels with Liquid Bismuth in Non-isothermal Systems. D. W. Dawe, G. W. Parry, G. W. Wilson. (*J.I.S.I.*, 1958, **190**, Nov., 271–276.) [This issue.]

Intergranular Corrosion Resistance of Austenitic Stainless Steels. A Ferric Sulphate-Sulphuric Acid Test. M. A. Streicher. (*A.S.T.M. Bull.*, 1958, April, 77–86). A test is described taking only half the time of the standard nitric acid test and as the corrosion products have no accelerating action, four specimens can be tested together and the same solution can be used for the whole 120 h exposure. The chromium carbide type of intergranular corrosion is revealed but not the sigma-phase type so that it is unsuitable for AISI 316 and 316L steels. Other etching tests are noted.

Weathering Resistance of Mild Steel with an Aluminium Coating. G. von der Dunk. (**Stahl Eisen*, 1958, **77**, Mar. 7, 294–298). A review, mainly from the literature, of the resistance of rolled Al coatings, mostly on wire, to steelworks atmospheres. Analyses and the results of examination are recorded, and the resistance of hot-dipped wires in rural atmospheres is also noted.

Determination of Intergranular Corrosion by Measuring the Internal Friction. M. A. Vedeneeva, A. V. Panov and N. D. Tomashov.

(*Zavods. Lab.*, 1957, **23**, (1), 64–67). Corrosion was induced with boiling H_2SO_4–CuSO4 solution and internal friction was measured with apparatus described and also electrical resistance and tensile strength. The former was the most sensitive. Susceptibilities to intergranulation were noted, and the effect of Ti in minimizing this was observed.

Corrosion Resistance of Austenitic Chrome–Nickel Steels. W. Katz. (*Draht*, 1957, **8**, Aug., 317–321). Following an outline of the principles of passivation and the causes of the breakdown of the passive layers in austenitic steels, the resistance to corrosive attack by the atmosphere, water, salts, acids, bases, and organic compounds of various austenitic steels is briefly summarized.—J. G. W.

The Formation of Chi Phase in a Mo-Bearing Stainless Steel. Chuang Yu-chih, Liu Chia-lo and Chang Ching-chung. (*Acta Met. Sin.*, 1957, **2**, (1), 81–97). [In Chinese]. In quenched 17/5/6 Cr–Ni–Mo steel, the decomposition of δ-ferrite gives chi-phase and austenite in the temp. range 700–1000° C, but Fe_3Mo_3C and $(Cr, Fe, Mo)_{23}C_6$ instead of chi in the range 500–700° C. The mechanism of chi formation varies with the reheating temp. Below 700° C martensite formation in the austenitic matrix accompanied carbide pptn. (30 references).—K. E. J.

Contribution to the Estimation of M_s from the Composition of Steels. J. A. Verčo. (*Acta Techn.*, 1957, **19**, (1–2), 193–198). [In English]. A quadratic expression is put forward connecting M_s with carbon content which is correct from 0·1–1·4% C. This is M_s (° C) = 520 − (360–500) C, where C is % carbon.

Nature of Isothermal Transformation between Room Temperature and Martensite Point in a Ball-Bearing Steel. Chang Pei-lin and Hsu Tsui-chang. (*Acta Met. Sin.*, 1956, **1**, No. 4, 347–366). [In Chinese]. The steel contained 1·43% Cr and 1·02% C. Isothermal transformation below the M point takes place by the martensite process entirely. Isothermal transformation of retained austenite takes place by growth of already-existing martensite, and is controlled by the tempering of martensite. The max. amount of transformation of retained austenite is brought about by isothermal retention at approx. 100° C for a long time. (20 references).—K. E. J.

On a Peculiar Structure in the Electron Micrograph of the Martensite in Fe–Ni Alloys. Z. Nishiyama and K. Shimizu. (*Metal Physics*, 1956, **2**, Mar., 71). [In Japanese].

Researches on the Martensite Transformation in Europe and U.S.A. Z. Nishiyama. (*Metal Physics*, 1955, **1**, Sept., 177–187). [In Japanese].

Studies on the Martensite Transformation in Iron–Nickel Alloys. II. Effects of Surface and Interface on "Schiebung" Transformation. T. Honma. (*Nippon Kinzoku*, 1957, **21**, Feb., 122–125). [In Japanese]. The "Schiebung" (glide) transformation shows a peculiar behaviour at the free surface of the crystal, the twin boundary and the interface between austenite and martensite previously formed by the "Umklapp" (collapse) transformation. The transformation temp. at the free surface and the twin boundary is usually higher by about 20° C than in the interior. The austenite-martensite interface is also a preferred site for "Schiebung" nucleation. These phenomena are explained in terms of free energy etc. **III. Coexisting Range of "Schiebung" and "Umklapp" Transformations.** (126–128). The two types of transformation can occur in specimens of the same nickel concentration by supercooling to low temperature. The highest temperature at which the "Umklapp" (collapse mechanism) transformation can occur is always lower than that of the "Schiebung" (glide mechanism), and is approx. − 10° C, irrespective of nickel content. **IV. Burst Phenomenon of Umklapp Process.** (Apr., 263–267). The Umklapp (collapse) transformation in Fe–Ni alloys containing 31% Ni is characterized by a burst phenomenon. High kinetic energy accompanying the process can bring a martensite leaf into a temp. range higher than that of the normal $\gamma \rightarrow \alpha$ temp. Values for the first burst temp. are scattered over a wide temp. range, and the amount of transformation increases with the degree of supercooling. (11 references).

A Thermodynamic Study of the Fe-C System. II. The Stable Iron–Graphite Equilibrium. Chou Yuan-hsi. (*Acta Met. Sin.*, 1956, **1**, (2), 209–220). [In Chinese]. The Fe–C equilibrium diagram is studied in the light of activity data previously derived. The solubility of graphite in ferrite is calculated using the free energy of cementite. Parameters affecting the Fe–C eutectic equilibrium are calculated. The δ-iron–austenite-liquid peritectic is best given at 1495° C with 0·58% C in the liquid. The solidus of the austenite-liquid field is slightly concave upwards. (20 references). **III. The Metastable Iron–Cementite Equilibrium.** ((3), 221–232). With the aid of activity data previously derived, the cementite eutectic temp. is found to be 1148° C, or 5° C below the graphite eutectic temp. The free energy of cementite is recalculated, and found to vary from that given by Darken and Gurry. The solubilities of cementite in austenite and liquid iron are calculated approximately in terms of temp. A revised Fe–C diagram including both stable and metastable equilibria is given. (22 references).

Studies on Cr–Mo–Fe Super-Heat-resisting Alloys. I. The Equilibrium Diagram of Cr–Mo–Fe System. S. Takeda and N. Yukawa. (*Nippon Kinzoku*, 1957, **21**, Apr., 275–279). [In Japanese]. The diagram was established from microscopical, thermal and X-ray results. The system has six phases: α (b.c.c.), γ (f.c.c.), ε (Fe_3Mo_2, hexagonal), σ (FeMo and FeCr, tetragonal), δ (high-temp. modification of Cr, f.c.c.) and χ ($Fe_{36}Mo_{10}Cr_{12}$, b.c.c.). Three invariant reactions are distinguished. (20 references).—K. E. J.

CORROSION

Bibliographic Survey of Corrosion 1954–1955. National Association of Corrosion Engineers. (1958, pp. 408). This is a compilation of abstracts classified into Testing, Phenomena, Corrosive Environments, Preventive Measures, Materials of Construction and Equipment, followed by a survey of industries in 10 groups. All appear to be drawn from familiar sources. Indexing is complete.

Anticorrosion Manual, 1958. (pp. 260). **General Corrosion Principles.** W. H. J. Vernon. (1–13). **Prevention of Corrosion.** (13–25). Reviews are given of current theory and practice. **Fretting Corrosion.** K. H. R. Wright. (26–34). A survey, previously published elsewhere. **Cathodic Protection.** K. A. Spencer and D. A. Lewis. (36–40, 42, 44, 46, 48–54). A review with 18 references. **Instrumentation.** H. M. Powell. (55–56). **Corrosion Resistant Materials.** (58–59). Tables of corrosive agents for choice of materials are given. **Aluminium and Aluminium Alloys.** F. C. Porter. (60–65). **Carbon and Graphite.** M. Buis. (67–68, 70–73). **Rubbers and Elastomers.** F. F. Jaray and A. E. Lever. (103–104, 106, 108). **Stainless Steels.** Firth-Vickers Ltd. (109–110). Resistance to various agents and applications in various industries are briefly indicated. **Zinc Coatings.** (135–138). **Preparatory Treatments: Flame Cleaning.** (147, 149). **Sodium Hydride Descaling.** (149–150). **Shot Blasting.** J. Carle. (151–152, 154). **Phosphating Methods.** 154, 156, 158). **Chelating Agents.** J. K. Aiken. (158, 160, 162, 164, 166, 168, 170). **Protective Paint Coatings.** (175–210). Sections on chlorinated rubber, epoxy coatings, lead paints, neoprene, vinyl resins, marine paints and other recent developments are presented. **Other Protective Coatings: Cadmium.** P. F. Norrish. (212, 214). **Tin.** S. C. Britton. (216, 218). **Metal Spraying.** (219–225). **Plastic Coatings.** G. A. Curson. (226–228). **Vitreous Enamel Coatings.** J. W. G. Pedder. (229–232). Most sections are reprinted from *Corrosion Prevention and Control*.

Methods for Increasing the Corrosion Resistance of Metal Alloys. N. D. Tomashov. (*Corrosion*, 1958, **14**, May, 229–236; similar to *Uspekhi Khimii*, 1955, **24**, (4), 453–470). A detailed study of the general laws governing the relation of corrosion-resistance to alloying additions and other topics such as heterogenity, passivity, oxygen diffusion and the effects of cathodic inclusions. A method of calculating relative passivity is given and in N/2 solutions of NaCl the order of reduction of degree of passivity is Ti, Al, Cr, Be, Mo, Mg, Ni, Co, Fe, Mn, Zn, Cd, Sn, Pb, Cu. Experimental tests of readily passivated systems are reported.

The Electrochemical Behaviour of a Passive Austenitic Chromium–Nickel Steel, a Chromium–Molybdenum Steel and of Passive Electrolytic Chromium in Boiling 20% Sulphuric Acid. C. Carius. (*Arch. Eisenhütt.*, 1958, **29**, July, 449–461). The effect of redox potentials on the phenomena occurring at a passivated electrode are discussed. The electrochemical and corrosion-chemical behaviour of passivated 18–8 steel, of a steel with 27% Cr and 2% Mo and of electrolytic Cr in boiling 20% sulphuric acid was studied. Stationary current/potential curves are presented and the results of gravimetric and coulometric analyses given.—T. G.

Corrosion Inhibitors in Automotive Coolant Media: Action of Polarization Characteristics of Steel. M. Levy. (*Ind. Eng. Chem.*, 1958, **50**, Apr., 657–662). Studies on inhibitors in ethylene glycol solution are reported.

Dicyclohexylammonium Nitrite as a Vapour-Phase Corrosion Inhibitor. S. Mihailovic. (*Zaštita Mat.*, 1955, **3**, June, 201–204). A review with 39 references.

The Importance of Corrosion Protection in the Metal Manufacturing Industry. S. Murgaški. (*Zaštita Mat.*, 1955, **3**, Oct., 334–336)! A review of the theory of corrosion and its prevention, and of the metal industry.

The Need for Setting up Services for Anti-corrosion Protection in Industry. F. Podbrežnik. (*Zaštita Mat.*, 1955, **3**, Oct., 337–338). Plating, metal spraying, packaging and storage are discussed.

The Temporary Protection of Metal Surfaces against Atmospheric Corrosion. N. Plavšic. (*Zaštita Mat.*, 1956, **4**, Mar., 69–75). The coating and inhibiting classes of protective agents are reviewed. The volatile inhibitors are given special attention.

The Protection of Steel Castings against Rusting. (*Journal of the B.S.C.R.A.*, 1958, April, reprint). Jute impregnated with sodium benzoate is used.

Protection Against Corrosion of Industrial Installations for Producing and Utilising Steam. R. Malicet. (*Corros et Anticorros.*, 1957, **5**, June, 174–184; July-Aug., 216–220). The first part deals with the corrosion taking place in a large variety of equipment for the production and utilization of steam for industrial purposes. Each type discussed is illustrated by photographs. The second part is concerned with methods of preventing such corrosion.—L. D. H.

Cathodic Protection for Earth-Buried Pipelines and Other Metal Structures. J. H. Bird and A. Z. Dimitroff. (*U.S. Bur. Mines Inf. Circ.*, 7827, 1958, Apr., pp. 27). General principles are discussed with special reference to earthing problems.

Corrosion Control in Tankers. A. Logan. (*Trans. Inst. Marine Eng.*, 1958, **60**, May, 153–174, discussion 175–194). A review of the problem with accounts of cathodic protection and of the use of epoxy tank coatings. The method of introducing inhibitors as mist is then outlined.

ANALYSIS

Contributions to Micromethods in Iron and Steel Research. H. Malissa. (**Mikrochimica Acta*, 1956, (1–6), 474–483). Isolation processes for micro-constituents are considered and physical and chemical methods of identification are reviewed.

Control of and Chemical Methods for the Determination of Common Elements in Steel. J. Rodriquez Garcia. (*First Nat. Congress Iron and Steel Industry [Mexico]*, 1955, 157–160). C, Mn, P, S and Si are briefly considered.

Analytical Instrumentation in the Control of Production in the Steel Industry. Q.B.A.M. Canales. (*First National Congress of the*

Iron and Steel Industry [*Mexico*], 1955, 299–303). Modern techniques, such as polarography, X-ray analysis and mass spectrography are discussed.

Determination of Carbon in Steel by Combustion and Conductimetry. R. Boulin. (*Chim. Anal.*, 1958, **40**, Mar., 72–76). The method may be applied to steels with as little as 0·01–0·1% C. The CO_2 is passed into NaOH solution and the change in conductivity determined.

Simplified Equipment for Determining Hydrogen in Steel. E. Marianeschi and R. Coletti. (*Met. Ital.*, 1957, **49**, Sept., 673–676). [In Italian]. The method described is a variant of the classical method of vacuum extraction at 650° C. Usually the H_2 extracted from the sample is compressed in a McLeod or Töpler apparatus and then measured. In the method proposed by the authors, the amount extracted from the sample is measured by the variation in pressure in the constant-volume extraction chamber when de-gassing the sample. A Pirani vacuum-meter with bridge circuit calibrated in ml of H_2 is used. Test results are given. (9 references).—M. D. J. B.

Hydrogen in Ferrosilicon. Analytical Methods and Average Hydrogen Contents of Some Commercial Grades. N. Christensen, K. Gjermundsen. (*J.I.S.I.*, 1958, **190**, Nov., 248–254.) [This issue.]

On a Source of Error in the Determination of Nitrogen in Special Steels by the Wet Method. R. Castro, J. Allemand and R. Poussardin. (*Anal. Chim. Acta*, 1957, **17**, Dec., 530–534). [In French]. Traces of nitrate in the acid and traces of Mo in the metal, even a few ten-thousandths, give high results without necessarily having any effect on the blank. A test with electrolytic Fe with a minute addition (~ 0·1%) of Mo is described.

Method of Determining Sulphur in Various Materials by Reduction with Zinc Vapour. M. S. Kurchatov. (*Zavods. Lab.*, 1957, **23**, (11), 1294–1296). [In Russian]. The method for sulphur determination described is based on its conversion into sulphate, the reduction of this to sulphide with zinc vapour, the decomposition of the sulphide with hydrochloric acid and the iodimetric determination of hydrogen sulphide. Results obtained by this, the Eschke and the metallic potassium methods for cokes and coals are compared, the special modifications necessary for fuels being stated. The proposed method is suitable for slags, ores, steels and other alloys, a determination takes 1·5 h.—S. K.

The Photometric Determination of Molybdenum in Zinc Cyanide Plating Solutions. J. P. Leftin. (*Plating*, 1957, **44**, Apr., 380–382). A thiocyanate-$SnCl_2$ method using Butyl Carbitol as colour stabilizer is described.

The Determination of Vanadium by the Spectrophotometric Method in Non-Aqueous Solution. D. Cozzi and G. Raspi. (*Anal. Chim. Acta*, 1957, **17**, Dec., 590–596). [In French]. The colour produced by acidified H_2O_2 in dioxan (red-yellow) and by subsequent addition of salicylaldehyde (violet-blue) is visible even with 1 μg/ml of V cannot be used quantitatively as any contained water hydrolyses the coloured products. The residue (dried at 100°) dissolved in HCl and H_2O_2 solution can be used if acetic anhydride, acetic acid and methyl salicylate are added. Co, Mo and W interfere.

Detection and Colorimetric Estimation of Vanadium by Tiron. B. Sarma. (*J. Sci. Ind. Res.*, 1957, **16B**, Dec., 569–570). Tiron is sodium catecholdisulphonate, it can be used for Fe, Ti, Ce, Nb and U. It gives a bluish-violet complex with quinquevalent V in weak acid solution, and the interference of the above ions can be suppressed.

Microanalytical Investigations in Iron Research. III. The Handling of Micrograms in Metallurgical Investigations. W. Koch, H. Malissa, and D. Ditges. (*Arch. Eisenhütt.* 1957, **28**, Dec., 785–794). A micromanipulator is described that can also be used for mechanical and chemical studies. The taking of samples by drilling and the preparation of polished microsections are described. The samples are divided by means of small needles, permanent and electromagnets, or pincettes, mounted in the manipulator. Microgram samples are prepared with an electrically-heated platinum wire, using borax and sodium peroxide.—L. J. L.

The Determination of Microgram and Sub-Microgram Amounts of Boron. I. Absorptiometric Determination Using Curcumin. G. S. Spicer and J. D. H. Strickland. (*Anal. Chim. Acta*, 1958, **18**, Mar., 231–239). The method is particularly intended for the examination of materials for nuclear piles.

Complexometric Method for the Determination of Calcium and Magnesium in Iron with the Use of Cation-Exchange Materials. O. V. Datsenko. (*Zavods. Lab.*, 1957, **23**, (11). 1296–1298). [In Russian]. In the method described the solution of the sample of iron ore is treated with citric acid at pH = 8, The complex anions thereby produced, which contain the Fe and Al, pass through a column of cation-exchange material while the Mg and Ca are retained. If Mn is present the solution is treated with 1% hydroxylamine hydrochloride to prevent oxidation. Results obtained for Kerch and two classes of Krivoi Rog ores and flue dust (1·16–6·64% Ca, 0·98–4·38% Mg) obtained by this and the standard methods are compared.—S. K.

The Determination of Trace Amounts of Lead and Bismuth in Cast Iron. R. C. Rooney. (*Analyst*, 1958, **83**, Feb., 83–88). A solution in HCl is extracted with *iso*butyl acetate and then diethyldithiocarbamate complexes are extracted with $CHCl_3$ from ammoniacal tartrate-cyanide solution. Cathode ray polarography then determines the two elements simultaneously. Purification of reagents is described.

The Photometric Determination of Magnesium in Spheroidal Graphite Cast Iron by means of Eriochrome Black T. H. Pohl. (*Giesserei, Tech. Wiss.*, 1957, (17), July, 939–941). The literature on this subject is first briefly reviewed (14 references). A photometric method for determining Mg in spheroidal graphite cast iron, is described, the content varying from 0·005% to 0·15%. The influence of impurity elements is also dealt with and the calibration of the apparatus described. Full details (including numerical examples) are given of the reagents used and the method of carrying out the analysis. Finally, the method is criticized and sources of error discussed.—R. J. W.

The Polarographic Analysis of Tin and Tin Alloy Plating Solutions. A. E. Knotowicz and J. V. Petrocelli. (*Plating*, 1957, **44**, June, 602–606).

An Investigation of Conditions of Separation of Zinc and Cobalt from Aluminium and Iron Using Radioactive Atoms for the Control of Separation. Yu. V. Morachevskii and Z. S. Bashun. (*Izvest. Akad. Nauk. Otdelenie Khim.*, 1956, (10), 1185–1196). [In Russian]. Conditions under which complete separation of Zn and Co from Al and Fe can be obtained by the precipitation of the two latter metals with (1) NH_3, (2) in the form of basic acetates and (3) pyridine was studied, using radioactive ^{65}Zn and ^{60}Co. It was found that, with the first reagent, complete separation of Zn and Co from Al and Fe is obtained only at pH 5·5–5·9 and in the presence of at least 5% of ammonium salt. The second method can be used only when Al is accompanied by a larger proportion of Fe. Precipitation of Fe and Al with pyridine gives very good separation from Zn and Co.—V. G.

Electro-Chromatography in the Separation of Ions. V. Separation of Third Group Metals. A. K. Majumdar and B. R. Singh. (*Anal. Chim. Acta*, 1958, **18**, Mar., 224–225).

The Separation of Difficultly Soluble Deposits from Highly Dilute Solutions. C. Mahr and H. Klamberg. (*Arch. Eisenhütt.* 1957, **28**, Dec., 795–800). A method developed from circular filter paper chromatography is described. The method is based on the assumption that precipitation is prevented by brine formation and that flocculation is due to the electrical charge in the paper capillaries. Quantitative separation of the deposit occurs on small areas of a paper filter. Microgram amounts of Fe, Co, Ni, Pb and Cd have been precipitated from concentrations down to 0·01 p.p.m.—L. J. L.

Recommended Methods for the Analysis of Trade Effluents Methods for the Determination of Phosphorus and Acid-soluble Sulphate. A.B.C.M.-S.A.C. Committee on Methods for the Analysis of Trade Effluents. (*Analyst*, 1958, **83**, Jan., 50–52).

Designation of Shapes and Sizes of Preformed Graphite Electrodes. (*Amer. Soc. Test. Mat. preprint*, 1957, (55), 66–70). The electrodes are for spectrographic purposes.

Spectral Analysis from a Single Working Curve by the Method of Independent Evaporation in a Chlorine Atmosphere. I. M. Kustanovich. (*Zavods. Lab.*, 1957, **23**, (11), 1332–1335). [In Russian]. The aim of the work described was to eliminate the influence of the structure and composition of the sample on the results of spectroscopic analysis. This was effected by converting the element always into the same molecular compound (chloride) with gaseous chlorine. The apparatus is described and test results with iron- and steel-making materials are given and discussed, including sinters, ores, refractories, ferro-alloys. Errors for the main elements are given and it is stated that the method can be made more convenient at the expense of accuracy by using chlorine-evolving reagents instead of gaseous chlorine.—S. K.

Oxygen Content of Mild and Ultra-Mild Rimming Steels. A. Decker. (*Meeting in Belgium and Luxembourg*, 1958, *Liège, preprint*, pp. 4). Contents of various types of steel, (Kaldo, Rotor, L.D. and O.H.), are given and related to carbon content.

Spectrophotometric Determination of Cobalt with *iso*Nitroso-Dimedon. W. van den Bossche and J. Hoste. (*Anal. Chim. Acta*, 1958, **18**, June, 564–568).

The Determination of Chromium, Manganese and Vanadium in Steel by Automatic Coulometry. A. Liberti and L. Ciavatta. (*Met. Ital.*, 1958, **50**, Feb., 50–52). The determination of Cr, Mn and V in steel can be done automatically by coulometry by the electrolytic reduction of these elements at their state of maximum oxidation. A current 100% efficient in the reduction process is obtained by using as electrolyte a solution of a ferric salt. The equivalence point is obtained by means of a rotating Pt microelectrode polarized at + 1 V. The diffusion current, due to ferrous ions, stops the generating current automatically by means of an electronic relay. The procedure described is applied to standard analytical methods of determination.—M. D. J. B.

The Absorptiometric Determination of Traces of Chromium in Nickel and Vanadium, of Vanadium in Chromium and of Nickel in Chromium and Vanadium. J. T. McAloren and G. F. Reynolds. (*Metallurgia*, 1958, **57**, Jan., 52–56). Various separations and determinations are discussed comparatively. Results obtained with recommended methods are discussed.—A. G.

Rapid Determination of Manganese in Ferromanganese and Tungsten Tool Steels. C. O. Ingamells and W. S. Bradshaw. (*Chem.-Analyst.*, 1958, **47**, Mar., 11–12). Perchloric acid oxidation and titration with ferrous ammonium sulphate is used.

The Determination of Molybdenum and Tungsten. IX. The Estimation of Tungsten as Barium Tungstate. Liang Shu-Chuan and Hsu Pao-Yun. (*Acta Chim. Sinica*, 1956, **22**, (2), 93–102). Precipitation should be made from boiling solution at pH 7·2–8·0 with a large excess of $BaCl_2$ (0·8–2·8 millimolar after precipitation). Ignition should be at 500–900° C, they apparently occlude several tenth milligrammes of water if dried at 105–150°. Exclusion of CO_2 is unnecessary. Of the alkali salts examined Na and NH_4 acetates interfere least and KCl the most. **X. On the Colorimetric Determination of Molybdenum with Potassium Ethyl Xanthate.** (3), 171–176). [In Chinese with English Summary]. Organic solvents for extraction of the complex are examined. Petroleum ether and toluene give a solution stable for at least 3 days, (benzene, 14 min. only).

Petroleum ether can be used only with very low Mo concentrations; 0–0·3 mg Mo obey Beer's law in toluene.

Dithizone as an Indicator for the Direct EDTA Titration of Various Metals. A. Celso Costa. (*Chem.-Analyst.*, 1958, **47**, June, 39–40). Ni and other metals can be determined.

Proposed Standard Method for Total-Chromium Determination. M. J. Taras, M. Lieber and W. A. Moore. (*J. Am. Water Works Assoc.*, 1958, **50**, June, 832–834). The permanganate-azide method is standardized for effluent analysis.

Determination of Manganese in Steels Containing Chromium and Vanadium. I. M. Issa, M. Hamdy and A. El Hadidy. (*Chem.-Analyst*, 1958, **47**, June, 44).

Contribution to the EDTA Titration of Molybdenum and Tungsten. J. Kinnunen and B. Wennerstrand. (*Chem.-Analyst*, 1958, **47**, June, 38).

Colorimetric Determination of Molybdenum in the Presence of Tungsten: Modified Mercaptoacetate Method. D. A. Otterson and J. W. Graab. (*Anal. Chem.*, 1958, **30**, July, 1282–1284). Other interfering elements are discussed and methods for their elimination are given.

The Determination of Tin and Tin-Iron Alloy Weights on Tinplate using the Kunze-Willey Method. F. Cooke and C. E. A. Shanahan. (*Metallurgia*, 1958, **57**, June, 321–326). Results from the Kunze-Willey method are in satisfactory agreement with those using chemical dissolution and weight loss provided a de-aerated electrolyte is used for precise work. The technique is ideal for examining differentially coated tinplate.

The Determination of Traces of Zinc in Ferrous Metals. J. Migeon. (*Chim. Anal.*, 1958, **40**, Aug., 287–292). A separation with dithizone followed by polarography is described.

Photocolorimetric Determination of Zirconium with Morin. H. Tůma and N. Tietz. (*Coll. Czechoslovak. Chem. Comm.*, 1958, **23**, Jan., 142–145). [In German].

Determination of Zinc Using Acridine. S. Wakamatsu. (*Nippon Kinzoku*, 1957, **21**, Apr., 213–217). [In Japanese]. A rapid, accurate gravimetric method has been determined for estimating zinc in iron, steel and ores of Fe and Al. Acridine and ammonium thiocyanate are used to ppt. Zn in the presence of Fe^{II}, Ni, Al, Cr, V^{IV}, Mn, Ca and Mg. With a slight change in procedure, Co and Mo can also be estimated.—K. E. J.

Colorimetric Determination of Tungsten in Scheelite Ores and Mill Products. R. P. Hope. (*Proc. Australasian I.M.M.*, 1958, Mar., 51–64). $SnCl_2$–NH_4CNS is used.

Semi-Quantitative Determination of Vanadium in Soils Using Ascending Paper Chromatography. H. Agrinier. (*Compt. Rend.*, 1958, No. 19, 2761–2763). Vanadium is extracted by ammonia and H_2O_2 and then developed with 8-hydroxyquinoline. Sensitivity is of the order of 1 p.p.m.—A. G.

Steelworks Analysis by Spectrographic Methods. D. Manterfield. (*Proc. Congress Modern Analytical Chemistry in Industry*, 1957, *St. Andrews*, [1958], 34–42). Analysis of bricks is included.

Metallurgical Absorptiometry and Spectrophotometry. E. C. Pigott. (*Murex Rev.*, 1958, **1**, (18), 528–540). Photocells, photoelectric absorptiometry and available instruments are reviewed spectrophotometric high-precision absorptiometry, absorption microscopy, absorptiometric chromatography, emission absorptiometry and micro-absorptiometry are surveyed.

Spectrochemical Analysis by Means of the D.C. Carbon Arc. N. W. H. Addink. (*Lab. N. V. Philip's Gloeilamp. Separaat*, 2419; *Applied Spectroscopy*, 1956, **10**, (3), 128–137; from *Spectrochimica Acta*, 1955, **7**, 45–59). Tables of empirically determined K-values are given for the constant temperature arc method.

The Quantometer. L. Kidman. (*ESC News*, 1958, **11**, spring, 10–13). The lay-out of an installation, a description of the parts and a brief account of its operation are given.

Spectrographic Determination of the Enriched Elements in the Surface layer of Steel. II. Determination of Tin with an Intermittent Arc and a Feussner Spark as Light Source. T. Takei. (*Nippon Kinzoku*, 1957, **21**, Feb., 117–121). [In Japanese]. Optimum conditions were established for analysing the tin content (0·02–30%) in the alloy layer. This wide range was divided into three parts, and discharge craters were kept small because of the high concentration gradient. In the range 0·02–0·4% Sn, an intermittent arc was used; above 3%, a Feussner spark.

Spectrographic Analysis of Some Ceramic Materials. J. F. Brown. (*Trans. Brit. Ceram. Soc.*, 1958, **57**, Apr., 218–231). Several methods are described including a solution technique. Methods for various impurities and for F are given.

X-Ray Fluorescent Analysis of Hard Metal and of Carbides for Hard Metal Manufacture. L. Peterson. (*Jernkonts. Ann.*, 1958, **142**, (4), 203–208). A method is described for the X-ray fluorescent analysis of hard metals and carbides, the samples being dissolved in HF–HNO_3 with the addition of tartaric acid; the analysing crystal was lithium fluoride. Ti, Ta, Nb, Zr, Co and Fe can be determined accurately and rapidly.—L. D. H.

Studies on Analytical Methods for Trace Elements in Metals using Radioactive Isotopes. I. Determination of Zirconium in Steel. H. Amano. (*Nippon Kinzoku*, 1957, **21**, Apr., 260–263). [In Japanese]. Several methods of analysis were investigated, ^{95}Zr being used as tracer. The best method was found to be that involving pptn. of zirconium phosphate after separation from iron by electrolysis using a mercury cathode. (17 references).

Open Hearth Analysis. A. Hegarty. (*Gas World*, 1958, **148**, Aug. 9, 236). Change over to natural gas firing required a control system which was installed. A brief account is given with flow sheet of the sampler and analysing and recording units.

Apparatus for Continuous Determination by Conductimetry of Traces of Carbon Monoxide in Gases. J. Waclawik. (*Chim. Anal.*, 1958, **40**, July, 247–249). The CO is oxidized to CO_2 with I_2O_5 which is absorbed in NaOH solution and the conductivity measured. Electrolytic regeneration of the solution is used.

Determination of Magnesium in Nodular Cast Iron. J. Bernal Nievas and L. Serrano Berges. (*Inst. Hierro Acero*, 1957, **10**, Oct-Dec., 389–401). Three methods are given. The first is a volumetric method based on the separation of iron by electrolysis with a H_2 cathode, precipitating the Mg as a methyl dioxinate dissolving it in hydrochloric acid and back-titrating with $KBrO_3$. The method is rapid. For greater accuracy, separation of Mn and the rest of the iron is proposed. (12 references). The second is an indirect colorimetric method. The methyl dioxinate precipitate is dissolved in acetic acid and an azo-colour developed with diazobenzene-sulphonic acid. The influence of pH, excess $NaNO_2$ and lapsed time between diazotising and alkalisation was also studied. The third is a direct colorimetric method. Most of the iron is separated by electrolysis, the rest and the Mn are removed with ammonium persulphate. The colour is developed with Eriochrome T black in a 11·7 pH buffer solution and measurements made on a Beckman DU spectrophotometer at 540 mμ. This method allows the use of 1 g samples with as little as 0·03% Mg.—P. S.

Carmine: Reagent for Boron. E. A. Johnson and E. J. Newman. (*Organic Chemical Reagents, Monograph No.* 34, 1958, pp. 3).

Determination of Acid-Soluble Aluminium in Steel. E. T. Saxer. (*Blast Furn. Steel Plant*, 1958, **46**, May, 489–492). Methods of determining acid-soluble Al are first discussed and then full details of a method developed by the author are described. Al is weighed as phosphate.—B. G. B.

A New Reagent for the Colorimetric Determination of Aluminium. H. Green. (*Metallurgia*, 1958, **57**, March, 157–158). In the method described aluminon is mixed with ferron in a strongly buffered solution. This produces a true coloured complex over a wide range of Al concentration without any precipitation.—A. G.

The Rapid Determination of Calcium and Magnesium Contents in Ores and Slags. G. Graue and A. Zöhler. (**Technical Materials Testing Establishment of Hüttenwerk Phoenix A.-G.*). The use of EDTA is described.

Rapid Determination of Magnesium in Basic Slag: Application of the Ethylenediaminetetra-acetic Acid Method. S. Wakamatsu. (**Japan Analyst.*, 1956, **5**, (5), 264–267).

Volumetric Determination of Sulphate in Solutions Containing Chromic Acid (Dead-Stop Titration). R. Weiner and E. Ney. (*Metalloberfläche*, 1957, **11**, Dec., 377–379). After reducing the chromic acid and buffering the solution with acetate, sulphate is precipitated with barium chromate, filtered off, and the chromate equivalent titrated iodometrically, using the "dead stop" method of indicating the end point.—L. D. H.

Recommended Methods for the Analysis of Trade Effluents: Methods for the Determination of Residual Chlorine, Cyanides and Thiocyanate, Fluoride, Formaldehyde and Sulphite and Thiosulphate. A.B.C.M.-S.A.C. Committee on Methods for the Analysis of Trade Effluents. (*Analyst*, 1958, **83**, Apr., 230–241).

The Use of Mono Gas Analysers. M. T. de Grandsaignes. (*Mét. Constr. Mécan.*, 1958, **90**, Mar., 177–179). A description is given of this gas analyser which is an automatic type of Orsat analyser. The use of the equipment in foundries is discussed.—B. G. B.

Some Observations on the Measurement of Spectral Line Intensities by Microphotometry of the Photographic Plate. D. R. Curry. (*Metallurgia* 1958, **57**, March, 162–163). Experimental work on comparative line intensity measurements is reported and discussed.—A. G.

INDUSTRIAL USES AND APPLICATIONS

Materials and Structures. S. V. Gardner. (*Inst. Gas. Eng. Publ.*, 522, 1958, May, pp. 77). Materials for construction of all items of gasworks plant are discussed, including steel. Special steels are briefly noted and tubular construction is discussed.

Steel in Relation to the British Railways Modernisation Plan. C. F. Ryan. (*Rail, Steel Topics*, 1958, **4**, Spring, 4–12).

Steel Fabrication Lighter than Aluminium. J. B. Teeter and R. Rohrberg. (*Metalw. Prod.*, 1958, **102**, May 30, 950–951). Anticing nozzles for aircraft in stainless steel are 50% lighter and 27% cheaper.

The Metallurgy of Materials for Haulage and Cage-Suspension Gear. R. Jeffrey and B. J. Nield. (*Trans. Inst. Min. Eng.*, 1958, **117**, July, 705–719, discussion 719–722; *Proc. S. Wales Inst. Eng.*, 1958, **73**, July, 76–90; discussion, 90–93). Properties of wrought iron chains and the advantages of the use of steel are reviewed. The compositions and heat-treatments of steels are considered and the effects of service conditions discussed. A Mn mild steel and other low alloy steels were finally selected.

Selection of Steels for Springs. H. J. Elmendorf. (*Met. Prog.*, 1958, **73**, Apr., 80–84). Properties and heat treatments for various kinds of steel wire are shown and spring properties are related to design. Suitable choice of wire can be made from these values and heat treatment conditions can also be derived.

Paper-Thin Steel Sheet Cuts Weight of Aircraft. (*Mat. Design Eng.*, 1958, **47**, April, 128–133). Design data are presented for structures in which corrugated stainless sheet 0·016–0·001 in. thick is spot welded to a thicker skin. The applications of the combination in aircraft structures are given.

P.H. Stainless for Hot Airframes. R. W. White. (*Met. Prog.*, 1958, **73**, June, 74–78). The fabrication and heat treatment of precipitation hardening stainless steels used in aircraft parts subject to aerodynamic and engine heating are discussed.

Construction of Railway Rolling Stock. (*Weld. Met. Fab.*, 1958, **26**, Apr., 146–149). The Islington workshops of South Australian Railways have recently raised production levels by the extensive use of semi-automatic submerged-arc welding coupled with accurate planning and scheduling. Details are given of two recent production programmes, as examples of the use made of this type of welding.—G. F.

Electrical Steels, How to Choose and Improve Them. B. A. Ruediger. (*Steel*, 1958, **142**, June 9, 116–121). Considerations governing the choice of silicon steels for electrical purposes are described. Commercial silicon steels are tabulated and various coatings are listed.

Relation Between the Electrical Manufacturing and Iron and Steel Industries. R. Fernández Niño. (*First Nat. Congress Iron and Steel Industry* [*Mexico*], 1955, 295–297).

The Use of Steel in the Construction of Cyclotrons and Synchrotrons. Jurczyk. (*Acier-Stahl-Steel*, 1957, **22**, Dec., 497–498). The author gives a short summary of the essential parts of cyclotrons and synchrotrons, referring particularly to the use of steel. Mild killed steel of low carbon content is employed (< 0 05%C) for the poles. Cu, Sn and As should be absent.—E. P.

The Atomic Age Throws a Challenge to Steel. E. A. Livingstone. (*Mét. Constr. Mécan.*, 1958, **90**, May, 361–369). The special properties demanded of steels for use in the atomic energy industry are discussed.—A. G.

Costs and Operation of Mining Supports of Steel and Light Metal. F. Spruth. (*Glückauf*, 1958, **94**, Feb. 15, 280–292).

HISTORICAL

The Earliest "Furnaces". F. Evans. (*Foundry Trade J.*, 1958, Apr. 24, 491). A hypothetical section of a bloomery, with notes on its operation and products, is given.

Round Oak Steel Works Centenary. (*Metallurgia*, 1958, **57**, Jan., 33–35). Historical development and present plant are described. Future plans include 2–50 ton electric melting furnaces, which will bring total annual ingot production to 500,000 tons.—A. G.

An Eighteenth Century Iron Foundry. T. R. Harris. (*Found. Trade J.*, 1958, **104**, May 15, 585). An account of the products of the foundry of John Harvey at Hayle, founded about 1778.

Contributions to the Research on Prehistoric and Early Iron Production in the Lahn–Dill District. K. Heymann. (*Stahl Eisen*, 1958, **78**, June 26, 906–910). The author reviews the results of various excavations in the Lahn–Dill district. Extensive remainders of early iron smelting were discovered near the village of Dietenhausen. A few miles to the north-west of this village a number of smithies were found which used Siegerland iron as was shown by chemical analysis.—T. G.

Réaumur and the Discovery of Malleable Cast Iron. H. Abrecht. (*Schweizerische Bauzeitung*, 1957, **75**, Oct. 12, pp. 3, reprint). [In German]. Historical.

Early Derbyshire Furnaces. K. J. Haycock. (*Brit. Steel*, 1958, **24**, Apr., 109). Brief details are given of the construction and method of operation of two early blast furnaces in Derbyshire. They date from 1780 and 1818 respectively.—G. F.

Some Anniversaries in Iron and Steel. M. Schofield. (*Edgar Allen News*, 1958, **37**, Aug., 183–184). Notes on W. Kelly, Mushet and Siemens.

A Milestone in Ferrous Metallurgy. M. Schofield. (*Iron Steel*, 1958, **31**, May, 173–174). About 250 years ago at the Coalbrookdale ironworks, the elder Abraham Darby was the first to use coal or coke in place of charcoal in iron smelting. The author discusses the importance and implications of this step.—G. F.

Inventions and Discoveries of the 19th Century [In Iron and Steel]. G. Grenier. (*Usine Nouv.*, 1958, Feb., 97–98).

Sweden's Historic Iron Industry. F. W. Robins. (*Edgar Allen News*, 1958, **37**, May, 111–112).

Data on the History of Metallurgy in Hungary. VI. Project for a Blast Furnace and Blowing Engine Dating from 1816. A. Schleicher. (*Acta Techn.*, 1958, **20**, (1–2), 137–142). [In English]. Corrections to dimensions previously estimated for a ruined blast-furnace, since restored, are given and a plan of the blowing engine has been found.

John Wilkinson Ironmaster (1728–1808). J. H. S. Green. (*Research*, 1958, **11**, July, 250–253). A short biography.

John Jordan, Man of Iron. M. W. Fishwick. (*Iron Worker*, 1957, **21**, Autumn, 1–8). An illustrated biography of an iron smelter in the Shenandoah valley.

Hadfields of Sheffield. E. B. Thomas. (*Steel Rev.*, 1958, July, 36–44). An illustrated historical account.

The History of Richard Thomas & Baldwins Ltd. (*Met. Bull.*, 1958, Apr. Special Issue, 27, 29, 31). With special reference to tinplate.

ECONOMICS AND STATISTICS

The Steel Industry Metals. Iron and Steel. J. H. Thompson. (*Min. J.*, 1958, May, Annual Review No., 30–41). A statistical review. **Manganese.** A. G. Thomson. (33). **Nickel.** (35, 37). **Chromium.** (37, 39). **Cobalt.** (39). **Tungsten.** (39–40). **Molybdenum.** (40–41). **Vanadium.** (41). **Niobium and Tantalum.** R. B. Dunfield. (41).

Evolution and Outlook in the Italian Iron and Steel Industry. (*Usine Nouv.*, 1958, Feb., 95–96). Statistical.

Cost Accounting Procedures—For the Grey Iron Foundry. A. E. Grover. (*Foundry*, 1957, **85**, Nov., 110–113; Dec., 94–97; **86**, Jan. 103–107; Feb. 89–91).

Raw Material Requirements of the Steel-Using Industries in Mexico. M. Aramburu. (*Second Nat. Congress Iron and Steel Industry* [*Mexico*], 1957, 43–59).

First National Congress of the Iron and Steel Industry [Mexico], 1955. The Growth of the Iron and Steel Industry in Mexico. M. G. Aramburu. (85–91). Figures are given showing the production, consumption and import of iron and steel from 1939 to 1955.

National Panorama of the Iron and Steel Industry. P. G. Quintanilla (*Second National Congress of the Iron and Steel Industry*, [*Mexico*], 1957, 351–359). Figures are given for Mexico. Ore reserves of 500 million tons (61% iron) are estimated. In 1955, coal production was over 1·3 million t; in 1956 pig iron production was 408,309 t, steel 878,892 t.

Problems of the Steel-Using Industries. Comité Coordinador de las Industrias Metálicas de la C.N.I.T. (*Second Nat. Congress Iron and Steel Industry* [*Mexico*], 1957, 341–346). Mainly output statistics.

MISCELLANEOUS

High Purity Manganese via Electrolysis. (*Chem. Eng.*, 1958, **65**, May 19, 136–139). A flow sheet with illustrations. The raw material is furnace slag.

BOOK NOTICES

BOIE, WERNER. "*Vom Brennstoff zum Rauchgas*," Feuerungstechnisches Rechnen mit Brennstoffkenngrössen und seine Vereinfachung mit Mitteln der Statistik. 11¾ × 8¼ in., pp. iv + 102. Illustrated. Leipzig, 1957: B. G. Teubner Verlagsgesellschaft.

This book is devoted to the straightforward calculations that commonly have to be done by the combustion engineer: air for combustion; the volume and composition of, and total heat in, flue gas; flame temperature; and so on. The author claims that his dimensionless coefficients, derived from a 'statistical' evaluation of many analyses, are new, are characteristics of groups of fuels, and give simple mathematical expressions.

Whereas the material at any point in the book is simple, the fuel technologist who wants to use it must have familiarized himself with what precedes, or will have to do much searching back, since a reference list of the many symbols used, with definitions is conspicuously lacking.

The book requires of the student that he shall first have mastered the stoichiometry and routine of combustion calculations by conventional methods; he would then have to acquire the further technique of understanding and applying the new procedures. For the practitioner the value of the author's methods is most likely to appear in saving time and effort when many calculations of the same kind have to be done for the same or similar fuels, and would then be revealed by assiduous comparative trials.

The fuel scientist may sometimes find himself unwilling to trust the author's 'statistically' derived coefficients. They involve approximations and the neglecting of minor features. The author concedes that they do not always suffice for precise investigations. The word 'statistically' is misleading, for it means merely 'from the average of many analyses available to the author'; and the word 'dimensionless' does not have the esoteric connotation that mathematicians usually associate with it.

However, the book shows what can be done in its field, and its treatment of the field is interesting. The author's characteristics might be as useful for classifying fuels as for doing calculations. The book is very well produced.—R. A. ACTON TAYLOR.

BROUGHTON, H. H. "*Electric Cranes. A Manual on the Design, Construction, Application and Operation of Electric Cranes.*" 3rd edition, pp. 556. Illustrated. London, 1958: E. & F. Spon Ltd. (Price £7 10s.).

The author in this third edition has endeavoured to condense the whole variety of machines that lift material under the simple heading of 'Cranes.' This he has done with wonderful effect.

Chapter 1, for instance, emphasizes the large variety of machines that come under this heading. The slight digression to winches and conveyors is useful. The technical side of the book is complete, but because the minimum thicknesses usually specified in the U.K. and the U.S.A. are greater than those on the Continent, it is worth noting under 'Factor of Safety' that very many structural and mechanical members are determined by minimum limiting thicknesses or sizes which provide a much higher factor of safety than pure design warrants.

There are, as there must be in such a book, certain details that one may criticize. For instance, referring to the anti-friction bearings versus journal bearings in Chapter 14, the author ignores the very important fact that anti-friction bearings maintain the correct gear centres, keeping gear wear to a minimum. This, of course, is not possible with journal bearings. In para. 13 the author refers to 'Track Rails, Wheels and Trucks'; the rails indicated are sound for crane use, but when it comes to joints and fastenings a conventional flat bottom rail is indicated which makes this paragraph rather dangerous, because such rails have been found, by experience, not to have sufficient strength in the web to withstand lateral surge, except in the case of machines with extremely low acceleration on the cross traverse. His description of the important subject of electric equipment on cranes is extremely complete, so much so that it could lead to some confusion as to the choice of the correct equipment for a given condition unless one has knowledge on the subject. In para. 26 certain minimum clearances for overhead travelling cranes are given. These should be treated with care because the side clearance, and more particularly the headroom, is dependent upon the height of lift

in relation to the elevation of the crane rail. These comments are not intended to be critical of the author. It is inevitable that, in such a monumental volume, there should be points that can be chosen at random for criticism.

A very interesting and useful bibliography is provided at the end of each chapter. This well-indexed book will undoubtedly be for many years a standard work on the subject of the choice and design of hoisting machines under the heading of 'Electric Cranes,' and as such it ought to be in the library of every engineering firm.—R. BARTRAM.

COMBUSTION INSTITUTE. "*Sixth Symposium (International) on Combustion.*" 10 × 7 in., pp. 943. Illustrated. London, 1957: Chapman and Hall. (Price £11 4*s.*).

This symposium was held at Yale, New Haven, Conn. in August 1956. It is made up of 37 papers on the structure and propagation of laminar flames, many of simple systems of theoretical interest only, 6 on the structure and propagation of turbulent flames, 8 on high speed reactions, 8 on flame stabilization in fast streams, 4 on instability in combustion chambers, 5 on ignition, 8 on the combustion of solid fuels, 11 on the combustion of explosive and solid propellants, 9 on evaporation and combustion of droplets and sprays, 9 on experimental and analytical techniques in combustion, and 20 on applications of combustion.

For the steel industry, and indeed for furnaces in general, there is little of direct interest, though the papers in the analytical section on pyrometry and similar topics might find some application.

The last section refers mainly to rocket and gas turbine conditions and, oddly enough, to fire extinguishing. Nevertheless such fundamental studies cannot but be valuable for the understanding of high temperature processes and especially for furnace design. Books of this type, excellent in themselves, raise in acute form the implied conflict between pure and applied science or between the scientist and the engineer (decided, apparently, in favour of the latter in the recent U.S. questionnaire of High School pupils !), presenting as they do authoritative and most important data in what is manifestly a border-line field for the industrial scientist to make use of if he can, but undoubtedly enlarging his understanding and stimulating his capacity for observation of the processes under his charge and failing, in simplicity and in their special assumptions rather than in complexity and intricacy of formulation, to be of direct reference to the industrial environment. It appears to be at the root of the misunderstanding that the problems of the one and the solutions of the other should be too complicated for the two to meet on common ground. When scientists in different fields can no longer communicate with one another it is of paramount importance that they should be accessible to the applied scientific workers in their specialities, otherwise their work will be largely vain.

"*Deep Drawing Steels. Proceedings of a Regional Meeting, March 18–19, 1957, Detroit.*" Sponsors: Detroit Section, A.I.M.E., Detroit Section, National Open Hearth Steel Committee. I.M.D. Special Report Series No. 6. 11 × 8½ in., pp. 93. Illustrated. New York, 1958: The Metallurgical Society of A.I.M.E. (Price $6.00).

The twelve papers in this Symposium are divided into three particular aspects: Process Metallurgy, Physical Metallurgy, and Fabrication.

The section on Process Metallurgy thoroughly discusses capped steel, the effects of ingot structure, segregation and hot-strip mill practice upon the surface quality and mechanical properties of cold-reduced rimmed steel, the effects of hot rolling practice on the metallurgical structure of the strip and the relative advantages, both metallurgical and operational, of annealing in multiple or single stock coil furnaces.

Strain ageing is a major problem with deep-drawing steels and the section on Physical Metallurgy is mainly concerned with this subject, except for one paper on methods of analysis. The paper by Wert and Keefer neatly summarizes the role of internal friction methods in the study of the interstitial elements carbon and nitrogen which play a leading role in ageing. The use of internal friction to measure the solubility of these elements and their diffusion rates in α-iron is extremely interesting. The authors have ably presented both past work and some interesting original results.

Control of Strain Ageing in Steels—Theory and Practice, is a very brief summary of the Cottrell-Bilby theory of the discontinuous yield point and the behaviour of interstitial carbon and nitrogen. The effect of ageing temperature and the need to tie-up nitrogen in order to produce a stabilized steel is briefly considered. Boron, vanadium, and aluminium are nitride-forming elements used for this purpose.

A third paper in this section deals with the properties of rimmed, non-ageing vanadium-treated steel. Although the results are original and have not been seen elsewhere by the reviewer, there is one severe criticism to be made of the author's criterion of ageing. The ageing index used consists of a measurement of the difference in tensile strength at room temperature and 400° F (204° C). The difference if positive is an indication of ageing. Non-ageing steels show a weaker tensile strength at the higher temperature and hence a negative index. There are two criticisms of this method: (*a*) 400° F (204° C) is too high a temperature for accelerated ageing as pointed out in the preceding paper, and (*b*) it is only confusing to introduce a new 'ageing index' concept when the classical Gensamer and Low ageing index has already been in use for so many years.

In the third section on Fabrication, the three papers deal with the methods used to test deep-drawing sheet for its applicability to particular press shop use rather than with the actual fabrication. (Press shop, Metallurgy or Mechanical Testing might have been a more appropriate title).

The production metallurgist from both steel sheet producer and consumer industries will find a great deal of meat within the 93 pages. Although much of the information has already been published elsewhere, it is convenient to have all this experience together in one booklet. However, for its size and the magazine type of binding, it is rather expensive when compared with British standards.—S. GARBER.

"*Dislocations and Mechanical Properties of Crystals.*" An International Conference held at Lake Placid, September 6–8, 1956. Editors: J. C. Fisher, W. G. Johnston, R. Thomson, T. Vreeland, jun. 9¼ × 6 in., pp. xiv + 634. Illustrated. New York; London, 1957: John Wiley & Sons, Inc.; Chapman & Hall, Limited. (Price 120*s.*).

This is the report of the proceedings of a small international invited conference sponsored by the General Electric Research Laboratory, Schenectady, and the U.S. Air Force Office of Scientific Research, Air Research and Development Command. The topics discussed are classified into eight main sub-divisions: Direct observations of dislocations; Deformation of pure single crystals; Work hardening and recovery; Alloy crystals, impurities, yield point phenomena; Dislocation damping and fatigue; Theory of dislocations; Whiskers and thin crystals; Radiation damage. The 42 papers are mostly reports by workers active in the different fields of recent experimental and theoretical advances. The most exciting developments are to be found in the first section which includes papers on the direct observation of dislocations by various methods of 'decoration', by chemical and thermal etching, and, in thin metal films, directly by the electron microscope. The book is indispensable reading for those wishing to follow the rapid development of our understanding of the mechanical properties and imperfections in crystals. The full reports of the discussions which followed the presentation of the papers are a specially valuable feature. It is a pity that the price is such that the book is likely to be consulted mainly through the library.

A few years ago it was still necessary to justify to many the attention paid to dislocation theory, even in fundamental studies of mechanical properties. Although such justification is now no longer required, and many ideas from the fundamental studies already advance our understanding of the complexities of metallurgical practice, readers of the Institute's *Journal* will be most interested in those developments which promise further insight into the problems associated with materials actually in use. Unquestionably, important developments reported here are the demonstration of precipitation on dislocations in transparent crystals and the direct observation of dislocations in thin metallic films by means of the electron microscope. Already extended to films of stainless steel, and elaborated by the use of moire patterns, it is essential that the thin film studies should embrace as soon as possible other alloys and metallurgical phenomena. The thin film technique requires the best electron microscopes now available, and these instruments, combined with the latest replica techniques on bulk material, also make possible detailed studies of the complex precipitation phenomena in alloys of practical importance. It is to be hoped that as these instruments come into wider use, this work will be rapidly extended.

B. A. BILBY.

EVERDELL, M. H. "*Fundamental Thermodynamics for Engineers.*" General Engineering Series. 8¼ × 5½ in., pp. xii + 201. Illustrated. London, 1958: The English Universities Press Ltd. (Price 25*s.*).

An attempt has been made (in this first volume of a series intended for engineering degree students) to provide the basic theory of thermodynamics in a fairly short and inexpensive book. A large number of students today have to study the fundamental aspects of subjects which are closely related and essential to their main interest, but not central to it. Professor Christopherson, the General Editor of the series, makes this point in recommending the book as covering "all the Thermodynamics required" for Civil or Electrical Engineers, although, naturally, mechanical engineers will have to read deeper and wider.

Not everyone will accept the restricted view of the subject nor, as Professor Christopherson recognises, will all agree about what parts of Thermodynamics are important, but within its defined field the book provides an adequate description of the meaning, scope, and use of the First and Second Laws of Thermodynamics and of the thermodynamic functions, internal energy, enthalpy and entropy. Each chapter concludes with a number of problems, many taken from examination papers, with numerical answers and finally, the principles laid down are applied to the problems of heat engines.—M. L. P.

"*The Gray Iron Castings Handbook, Including Data on Gray Ductile (Nodular), White and High Alloy Irons.*" Edited by Charles F. Walton. 9½ × 6 in., pp. viii + 607. Illustrated. Cleveland, 1958: Gray Iron Founders' Society Inc.

This book, produced by the Gray Iron Founders' Society, Inc., Cleveland, Ohio, is an example of the recent trend towards joint authorship by a group of experts rather than by a single author.

As a protagonist for cast iron, the reviewer must applaud the Society for their treatment of chapter 2 wherein the many advantages of grey-iron castings to the engineer are skilfully pointed out to the reader and perhaps the authors may be excused for not directly mentioning many of the disadvantages. Useful data for

the design engineer and precise information for the purchasing officer are contained in chapters 3 and 4. Chapter 5 forms a larger part of the book and is divided into three parts. About 100 pages are devoted to the properties of cast iron and are treated in a manner which will commend itself to the engineer. Useful tables and graphs from 253 leading papers published during the last 20 years form a very valuable collection of design data. Part II of this chapter gives a very well balanced account of the potentialities of ductile (nodular) iron and some indication of its present importance. In the latter part of chapter 5 is introduced alloy cast irons and a short section on white irons, which perhaps may appear to be surprising in a book with this title. Chapter 6 on heat treatment deals mainly with heat treatment for stress relief, hardening and softening purposes. Welding as applied to cast iron has often been regarded as a last resort not to be attempted unless all hope of a repair by other methods has been abandoned. The authors of chapter 7 show that this is not the modern view and as a result of the increased understanding of the processes already in use, the employment of grey iron in welded assemblies is expanding. Chapter 8 deals with machining and grinding which commonly constitute the major item of production costs even with castings, which are always nearer the finished shape than steel billets or forgings. Tool requirements and machining speeds are placed on record for reference purposes. In the final chapter of 60 pages, methods of applying protective coatings are reviewed in fair detail including flame spray coatings of stainless steel, aluminium, hard facing materials, etc., on grey-iron rolls. Vitreous enamelling and the newer plastic finishes are perhaps rather inadequately treated. A glossary of technical terms and a short index complete this outstanding book.—R. V. Riley.

Himus, Godfrey W. "*The Elements of Fuel Technology.*" 2nd edition. 10 × 6 in., pp. xvi + 476. Illustrated. London, 1958: Leonard Hill (Books) Ltd. (Price 45*s.*).

The first edition of this excellent book, which was published in 1947, met with such success that it was out of print by 1952. In preparing the second edition, the author has kept generally to the original logical arrangement and has revised the text with the object of bringing it up-to-date. Though he has succeeded in including important recent information in most of the chapters, there are parts in which later information might well have been given. For example, the latest figures in Table I for the outputs of coals and petroleum for the world and for several countries are for the year 1951. In Table III the latest figure for the proved reserves of petroleum is only 7266 million tons, which was the accepted figure well over ten years ago. By the end of 1955 the proved reserves of petroleum were estimated at 26,000 million tons and the figure now accepted is more than 30,000 million tons, the increases being due to the discovery of new reserves at a rate greater than that of production.

More than two-thirds of the text is devoted to the origin, classification, and preparation of coals and to the utilization of coals for direct heating, steam raising, and the production of coke and various fuel gases. The theoretical aspects and the practice are well described and illustrated. Two chapters are devoted entirely to liquid fuels and one chapter to gaseous fuels and their combustion. The last chapter on domestic heating is particularly well balanced in relation to conditions in this country.

Dr. Himus has produced one of the best text books in this field, of value not only to students but to all seriously interested in fuels and their efficient utilization.

A. Parker

Johnson, Carl G., and William R. Weeks. "*Metallurgy.*" Fourth Edition. 8¼ × 5¼ in. pp. x + 454. Illustrated. Chicago, London, 1957: American Technical Society; The Technical Press Ltd. (Price 46*s.*).

This is a new edition of a general textbook on metallurgy which aims at providing an introduction to the subject, both for students and also for the practical man in industry. This new edition differs from the previous one mainly in the incorporation of a chapter on the special metals, titanium, zirconium, indium, and vanadium.

Whilst publishers apparently still consider that there is a need for a general introductory textbook covering the whole field of metallurgy it is obviously becoming much more difficult to compress the material into the few hundred pages available. Each new book of this type therefore must be judged by the way in which it covers this wide field and perhaps also by the way in which it acts as an introduction to further reading.

In this book the authors divide the subject into chemical and physical metallurgy and they rightly conclude that since more people are interested in the working and fabrication of metals the book should mainly deal with physical metallurgy. Extraction processes are therefore dealt with in two chapters. However, physical metallurgy in so far as this covers the structure of metals and the theory of alloys is covered in a further two chapters leaving the remainder of the book to deal with the "engineering properties of steel and non-ferrous alloys." In fact, this would probably have been a more descriptive title for the book.

There is a chapter on the testing of steel covering both mechanical testing and microscopical examination. There is also a chapter on the shape and forming of metals. After this, separate chapters deal with the structural metallurgy and mechanical properties of bearing metals, aluminium magnesium, and copper alloys. Five chapters are then devoted to steel and one to cast iron. One chapter on powder metallurgy and the new chapter on the special metals complete this book.

It often seems that there are too many general introductory textbooks on metallurgy and it becomes difficult to choose between them. This book covers a wide field quite effectively and therefore can be recommended. However, British readers should be warned that throughout the book the alloys and specifications referred to are American which limits its use as a reference book.—K. J. Irvine.

Pearson, W. B. "*A Handbook of Lattice Spacings and Structures of Metals and Alloys.*" International Series of Monographs on Metal Physics and Physical Metallurgy. Editor: G. V. Raynor. Volume 4. 10 × 6¼ in., pp. x + 1044. Illustrated. London, New York, 1958: Pergamon Press. (Price £13 2*s.* 6*d.*).

The need for a reference work containing information about the structures and lattice spacings of all binary and ternary alloys that have been examined has been widely felt by physicists, metallurgists and others concerned with alloys, and this first comprehensive compilation on the subject should fulfil their needs. The literature to the end of 1955 has been surveyed in full and the more accessible journals followed through 1956. In all there are about 6000 references. The accuracy of transcription is high. Where available, brief but penetrating comments are given on the constitution of the alloy systems (no equilibrium diagrams), the purity and heat treatments of specimens. Information about X-ray techniques is sometimes given and sometimes not. Published data on the variation of parameters with temperature are included. Many parameter/composition variations are illustrated graphically. These diagrams are clear and helpful though omission of 'weight' in the labelling of the abscissa of Fig. 180 is regrettable. The unfortunate heading of Chapter XII, 'Alphabetical Index of Work on Borides, Carbides, Hydrides, Nitrides and Binary Oxides' would not lead one to seek here data on the primary solid solutions of boron, carbon, etc. in metals. This initial difficulty in use is aggravated because the alphabetical arrangement followed in part II (959 pages) is considered a good reason for this part to be given only an outline index.

The first 174 pages of Part II gives atomic parameters and other information on the 'Strukturbericht' types, and tabulates crystallographic data on elements and compounds classified alphabetically and by structure type. These tables outdate other compilations.

Part I (72 pages) deals with units, the Debye–Scherrer method of parameter measurement and the lattice spacing method of determining phase boundaries. It summarizes various aspects of alloy theory, and discusses the relationship between lattice spacings and certain physical properties of alloys. The wide interests and balanced viewpoint of the author are apparent but the elementary and the advanced, the experimental and the theoretical are intermingled. No discussion is given of X-ray spectrometer methods which are becoming widely used for parameter determination. Much of the ground in the 'Theory of Alloys' Chapter is covered in *The Structure of Metals and Alloys* (Institute of Metals Monograph and Report Series No. 1, Third Edition 1954) by Hume-Rothery and Raynor. The chapter on the relationship of lattice spacings to magnetic properties reads like a review article.

The book is well produced and the binding should withstand extensive use. Despite the high price, though this is not extortionate, no metallurgical or metal physics library should be without it.

P. C. L. Pfeil.

Tödt, Fritz. "*Metallkorrosion. Allgemeines, Messung und Verhütung.*" 2 Auflage. 7½ × 5½ in., pp. [viii] + 122. Illustrated. Berlin, 1958: Walter de Gruyter & Co. (Price DM. 16.20).

Much of the loss in time and material due to corrosion could be avoided if designers, constructors and users of equipment had a greater awareness of the basic principles of the causes of corrosion and its prevention. There are many excellent texts for the corrosion engineer and the student of corrosion, but there is a real need for a short simple account of the main features of corrosion which would be of general use to engineers. This book goes far to fulfill that need.

This book is a second edition of a book first published 15 years ago and not, to my knowledge, previously available in this country. It is divided into three main sections referred to in the title. The first general section deals with the practical importance of corrosion, the fundamental significance of corrosion and theoretical considerations, the forms in which corrosion appears, the recognition of the controlling factors, and the influence of time on the course of corrosion. This is done without the use of chemical or mathematical equations which should help to popularize this book with people whose knowledge in these subjects is limited.

The second part deals with the measurement of corrosion and includes methods of testing in stagnant and moving liquids at room and elevated temperatures and testing in both natural and laboratory atmospheres. The various techniques used in assessing corrosion are enumerated. There exists a very wide variety of metals and environments and in many cases data is not available for the behaviour of a given combination, but much can be done by engineer or chemist to remedy this by means of relatively simple tests. This section should, therefore, be of particular interest on this account.

The third section comprising half the book, deals with corrosion prevention.

This is subdivided according to the different metals and alloys and hence the general applicability and methods available for the reduction of corrosion of any particular material can be readily ascertained.

The references are kept to a minimum and are mostly to standard texts on corrosion. The presentation of the book is very attractive and despite the obvious compression required the style is very fluent and readable. However, like many books these days, the price at about 30*s*. is rather high.—G. BUTLER.

TOURRET, R. "*The Performance of Metal-Cutting Tools.*" 10 × 6½ in., pp. xii + 184. Illustrated. London, 1958: Butterworths Scientific Publications. (Price £2 10*s*.).

This is the most comprehensive summary of published research work on metal cutting tools that has yet appeared. Not the least of its virtues is a massive list of references. Most of these are the sources of the data from which the book is compiled. Reference to the original research work is therefore facilitated.

This book is suited for those daily concerned with cutting tools who wish occasionally to refresh their memory on the published research. To such this book is invaluable. It is clear and concise. Further, its English is a delight to those who believe that it is the duty of technicians to express themselves clearly.

Despite its virtues, this book has its weaknesses. Some of these are inherent in the nature of the subject. Research work proceeds at such a pace that this book will need a companion volume in only a few years time if the subject matter is to be kept up to date. Part of the knowledge available has been thoroughly covered by published research work. In other areas, knowledge lies thin upon the ground. The book, therefore, in summarizing this, is bound to be 'bitty'.

It is surprising to find such little emphasis on the subject of the cutting tool material. This is perhaps the most serious weakness of the book. There is little, if any, mention of ceramic, or cemented oxide as a cutting tool material. There are references to published work on the effects of tool composition on performance. There does not appear to be any indication of the importance of grain-size to carbide cutting-tool performance. The effect of tantalum carbide is only very briefly mentioned and the effect of carbides of other metals appears to have had no mention at all.

This book should be on the shelf of every senior technician connected with the appropriate type of work. It has fulfilled a long felt want. It now needs to be followed by other books, just as authoritative, translating published research work into forms easily assimilated by serious students just beginning to specialize in this, a most important field of knowledge.—L. FINE.

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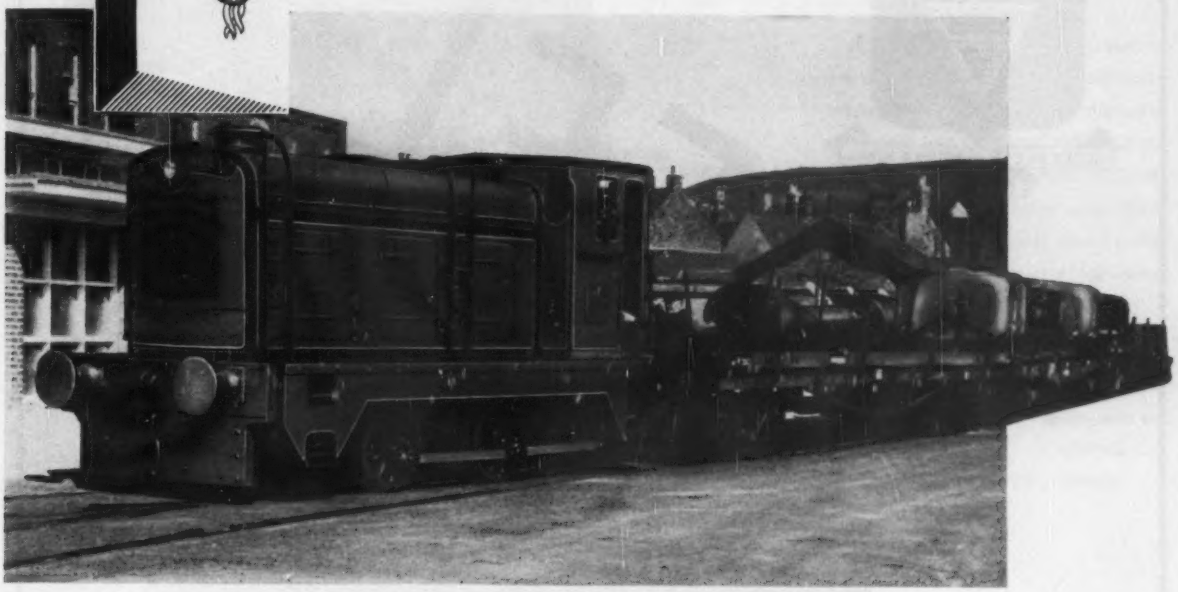

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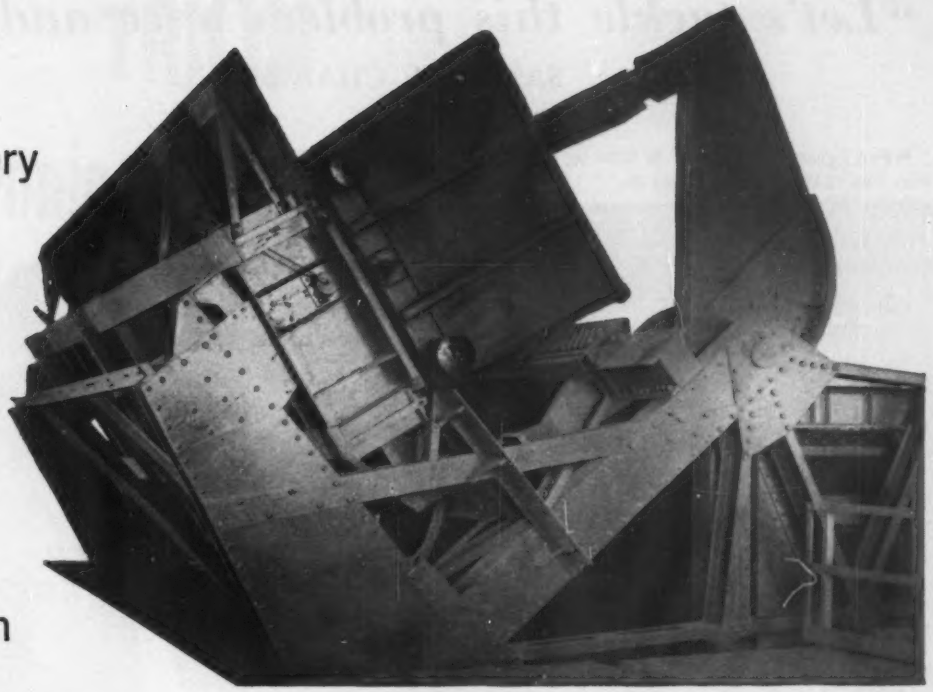

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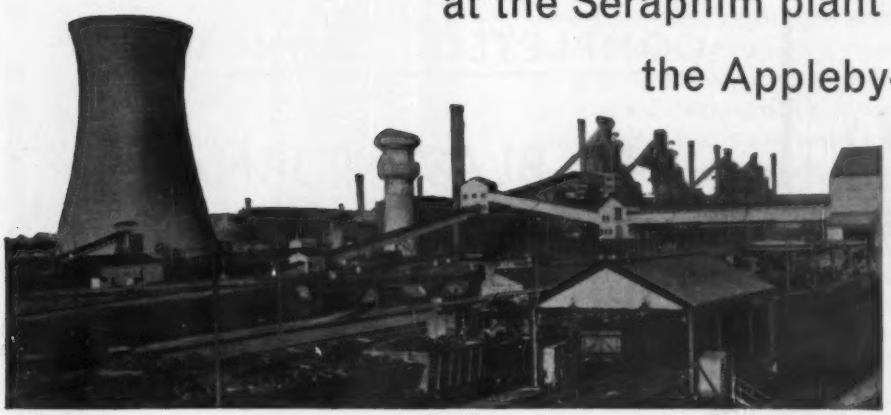

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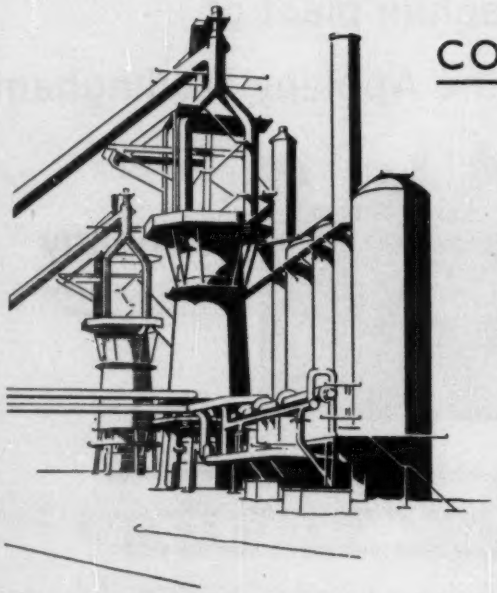

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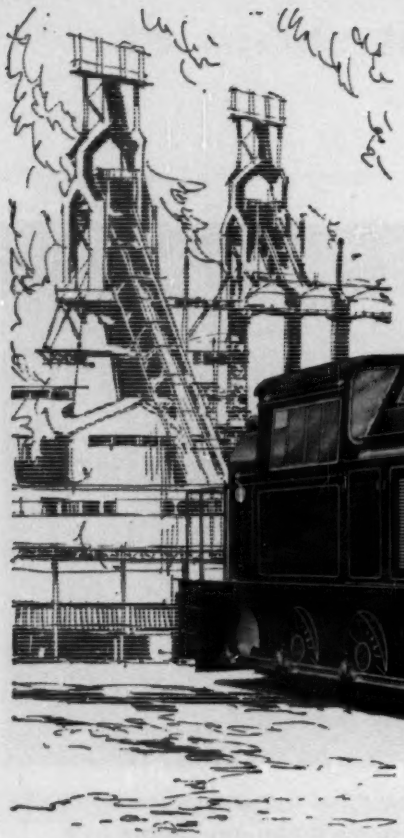

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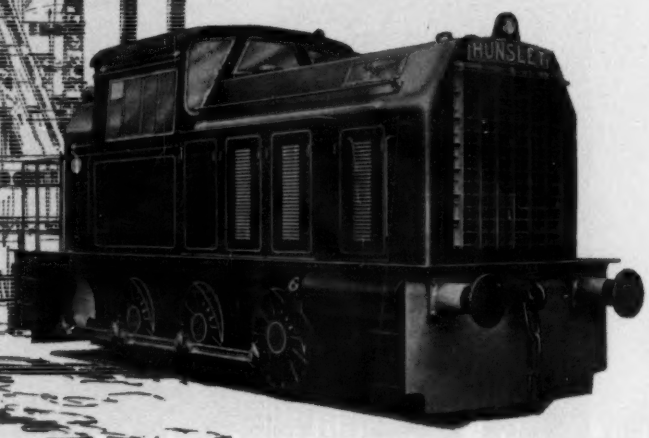

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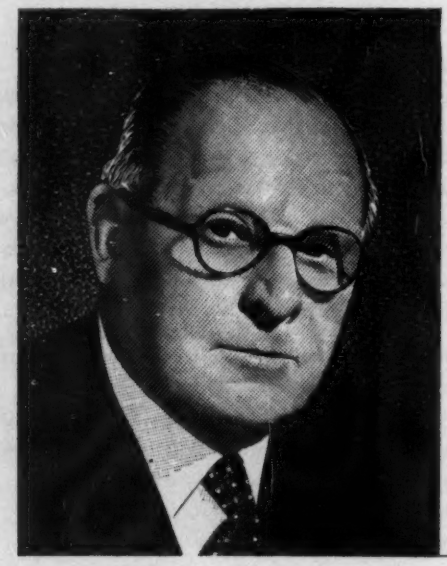

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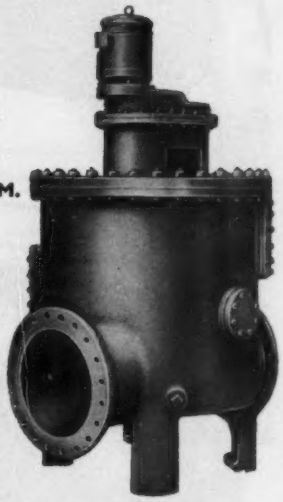

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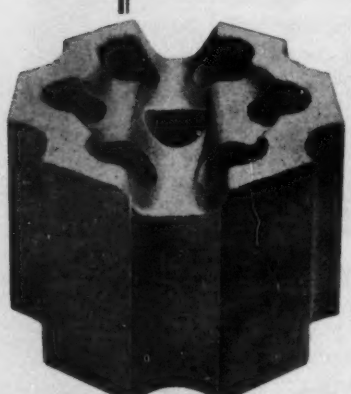

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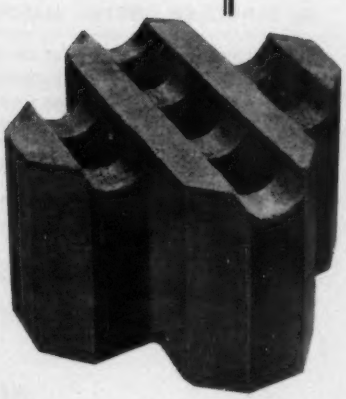

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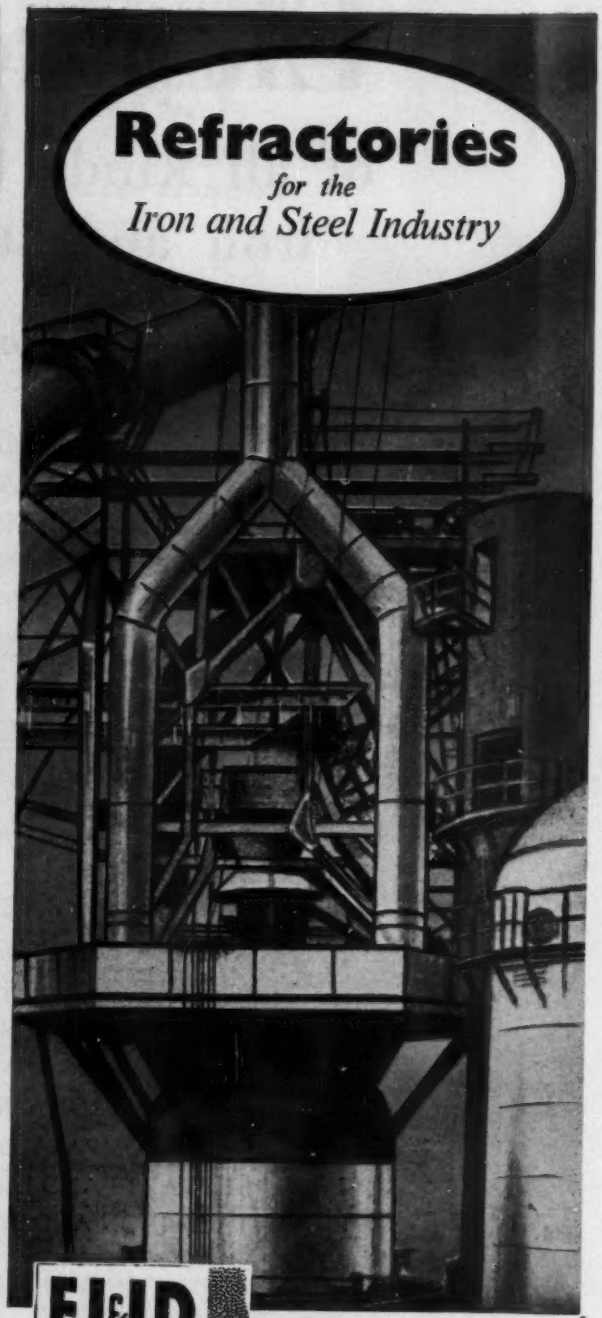

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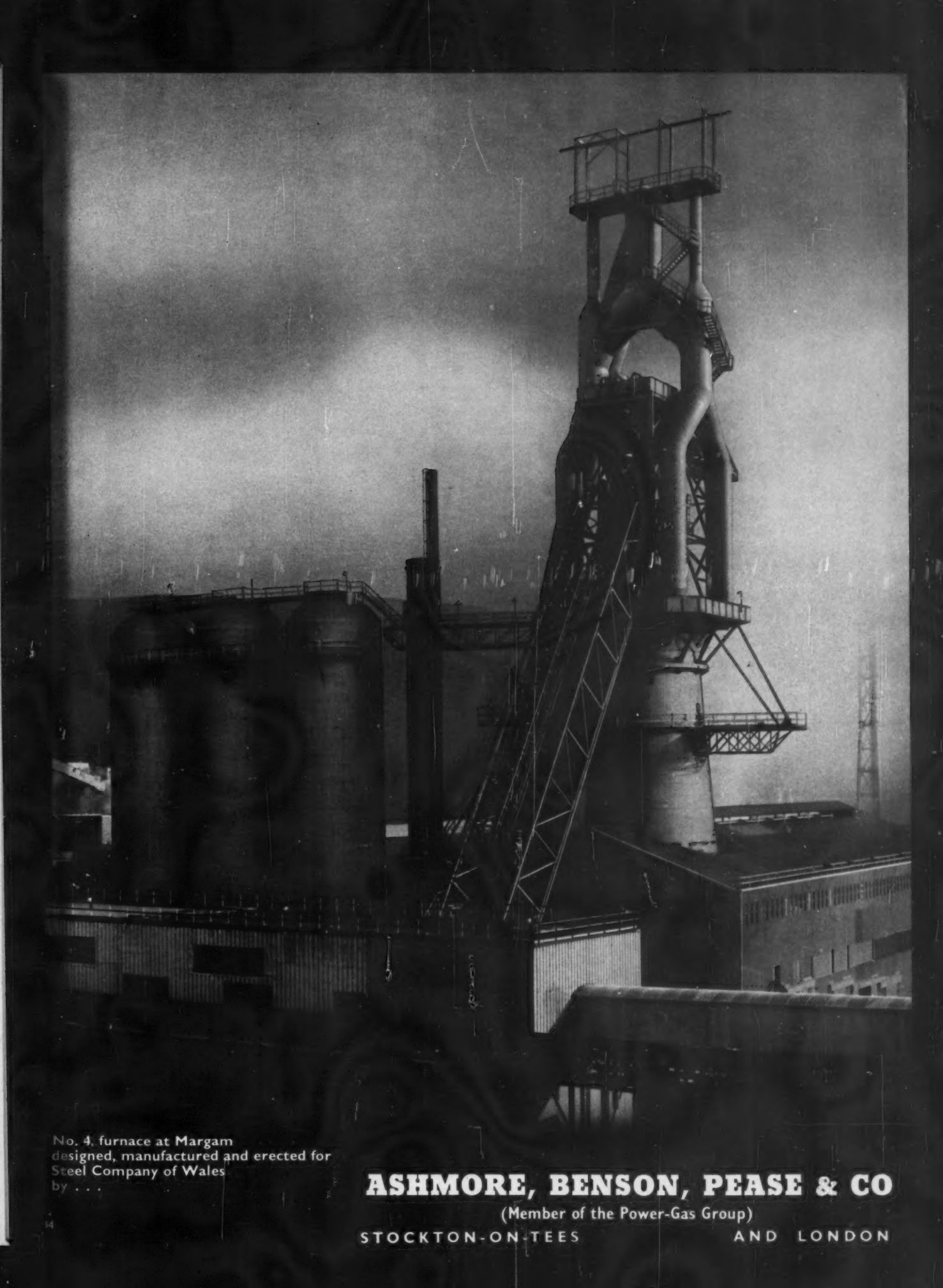

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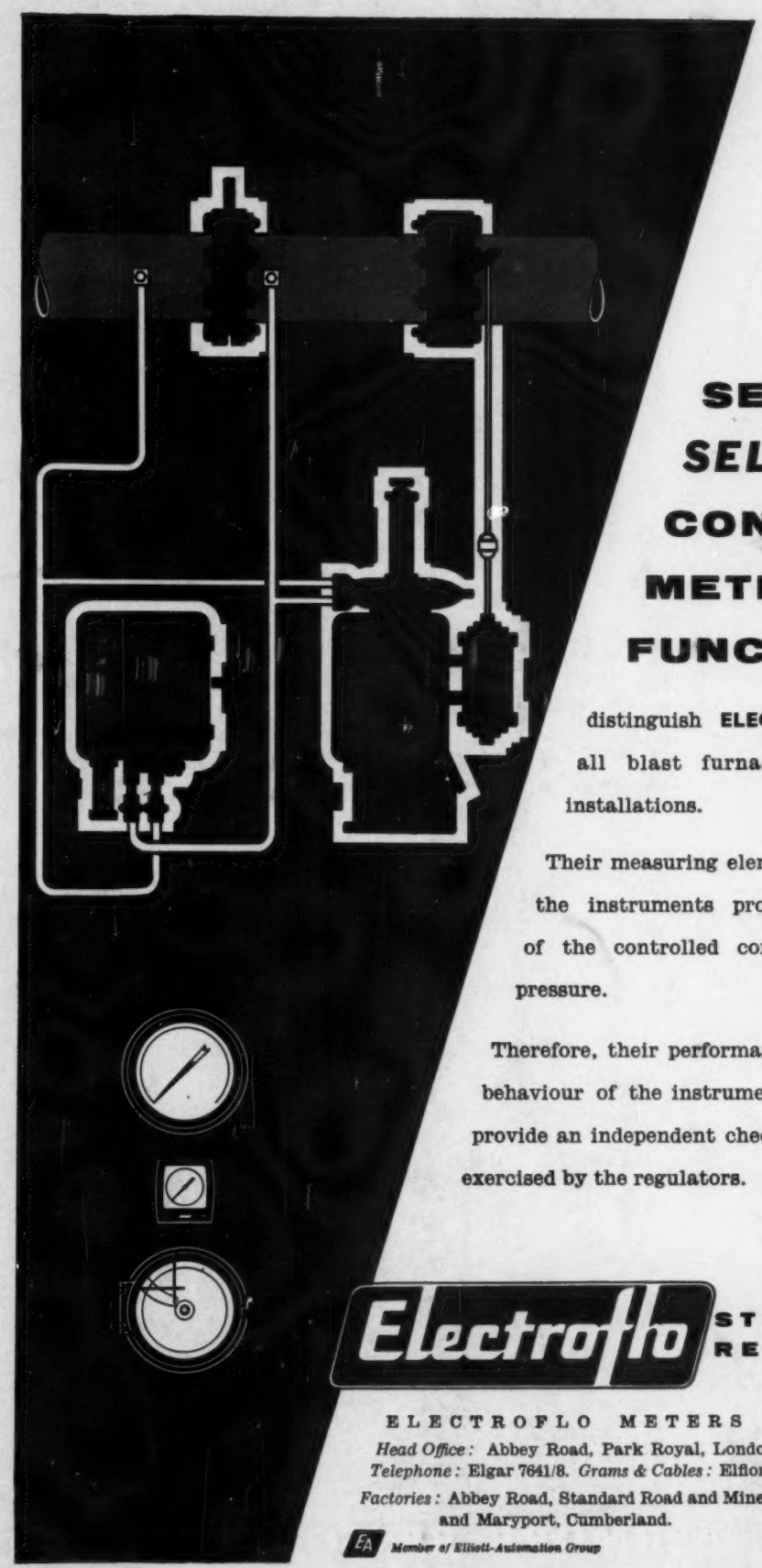

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